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Metals and Alloys

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Is there a Skeptic

We Quote These Facts

1. "Using U·S·S Controlled Steels, these tin snips have been easier to forge. In addition, the drilling operation has been more satisfactory, and heat treatment has been more uniform. Better tapping, grinding and polishing practice has been possible."

2. "U·S·S Controlled Steels have increased the physical properties of our formed and heat treated chain by at least 25%. We find the steel for our requirements is now at a very high standard of quality and of dependable uniformity."

3. "Since using U·S·S Controlled Steels there has been far more uniformity in all the processing operations, and many troubles have been eliminated. Better cutting of teeth, and hardening with less breakage and warpage are characteristics of the use of U·S·S Controlled Steels."

4. "In making Stillson wrenches performance has proved U·S·S Controlled Steels far superior for uniformity—has given these wrenches the ability to meet practically double the strength requirements. Made of U·S·S Controlled Steels, side cutting pliers have shown greater strength and longer life. Response to case carburizing and heat treatment has been excellent."

5. "U·S·S Controlled Steels, used for connecting rods, have proved advantageous in many ways. Ease of forging, uniformity in heat treatment and well-filled sections are characteristic. Also, freedom from trimmer cracks and quench cracks was noticeable, as well as marked toughness after heat treatment."

6. "With U·S·S Controlled Steels reruns on ring gears have been considerably reduced, the majority now being rerun for size only. Satisfactory performance has been noted with both fine and coarse grain steel, grade selected according to definite requirements."

UNITED

Tut-tut, Times!

The Feb. 12, 1939, Sunday *New York Times* carried an article "Light Alloy for Planes" by John McIntosh in which the metallurgist of a company exploiting beryllium is quoted, with apparent approval, as saying that it is proposed to offer Be metal at $11 per lb., that beryllium copper shows "high resistance to crystallization" in fatigue, that Be has practically no thermal expansion, and that it is resistant to heat of "more than 5000 deg. F."

Our copy of the A.S.M. Metals Handbook tells us that the coefficient of thermal expansion of Be is about the same as that of iron, that Be melts at 2462 deg. F. and boils at 2732 deg. F. We don't need to look up anything to form our own opinion of a metallurgist or a metallurgical editor who still refers to a fatigue failure as "crystallization."

Sound statements of metallurgical matters that will inform and interest the layman are good publicity for the metallurgical industries in general, but when a paper of the standing of the *Times* gets sucked in by such promoter-like statements and doesn't check them up for accuracy before publication, something is wrong somewhere. A leaf might well be taken from *Fortune's* book and the *Times* fined $5 for each misstatement of fact.

The only funny part of it is that in the article just below this one a crack is taken at aircraft makers for over-estimating performance and the comment made that "misleading advertising can be dangerous, often fatal."

It is perhaps of no moment that in a single instance the *Times* has "been had," as our English friends would put it, but that no mechanism has been set up whereby technical statements are checked up so that the layman is not misled, is a serious situation.—H.W.G.

Unit Metallurgical Operations

The mental processes of man are among the most peculiar and the most fascinating of natural phenomena. One of the most interesting phases of this is the difficulty, frequently experienced, in solving a simple problem because the basic pattern or mosaic of its structure, although visible, is not at first evident. When through accident or outside intervention the parts that make up the whole suddenly appear in proper orientation and relation to each other, we are abashed that something so simple of solution had puzzled us at all. Metallurgical engineering and many of its problems might be a much more simplified whole if we could learn to see the interrelation of its parts more accurately.

Our worthy brethren, the hard working chemical engineers, have found it convenient and advantageous to break down the processes by which they make their products into a series of unit operations. In Dr. A. D. Little's own words "any chemical process may be resolved into a co-ordinate series of unit operations, as pulverizing, drying, roasting, crystallizing, filtering, evaporation, electrolyzing, and others". This method of thinking has not only aided the chemical engineering student in his early understanding of operations in the process industries, but has remarkably clarified the problems of the practicing chemical engineer, for it has given him a thoroughly systematic basis for laying out new processes and a rational method for estimating costs. Because equipment for doing the same unit operation is fundamentally the same whatever the end product, designing in terms of unit operations opens up for the individual chemical engineer a vast fund of prior experience on other processes that can apply, with suitable adjustment, to his own.

How often do the makers, treaters, fabricators and users of metals employ this sort of resolution in attacking large scale metallurgical problems? In some fields, continually—in others, seldom. The principle has clearly been applied in those co-ordinated metallurgical operations that are closest to chemical engineering, *e.g.* the extraction and reduction of metals from their ores and their conversion into raw material—castings, sheet, plate, bar, rod, wire, etc.—for general industry. In general industry itself, the principle has been applied with notable success in many of the large mass production plants—particularly in the automotive field—to those operations that come under the general headings of working, heating and heat treatment, joining, and finishing. There are still, however, countless plants making products of metal that might profit by studying the analogy between chemical engineering control in a du Pont plant and metallurgical engineering control in a Ford

(Continued on page 89)

Properties of Some Free-Machining Lead Steels

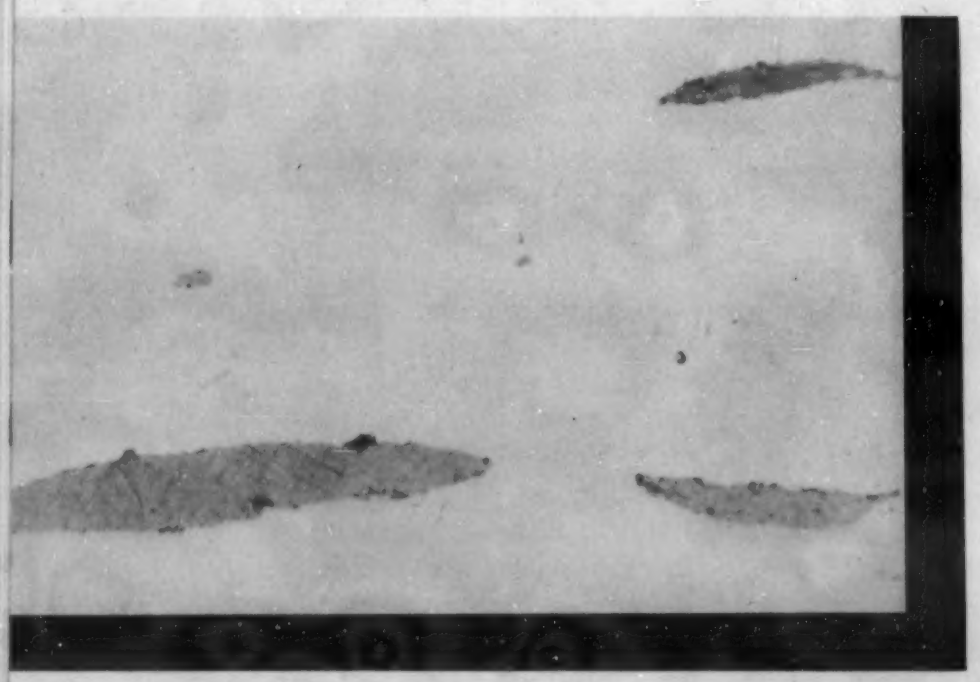
by J. H. Nead, C. E. Sims and O. E. Harder

Mr. Nead is chief metallurgist, Inland Steel Co., Chicago, and Messrs. Sims and Harder are of the staff of Battelle Memorial Institute, Columbus, O.

Ever since the first announcement of the new free-machining open-hearth lead-bearing steel—"Ledloy"—was made by the Inland Steel Co., Chicago, there has been keen interest in its metallurgy and properties. We are finally able to publish an authoritative article by some of those intimately associated with the development of this striking product.

As the authors say: Lead is but an infant in the kindergarten of the alloying elements. As regards screw stock, it is quite a lusty infant. As it grows and passes up through the grades, it may be graduated into an important alloying element in a much wider field—improvement in machinability of strong, heat-treated steels.—The Editors.

Fig. 1. Commercial Screw Stock Lead Free. Unetched. 1000 X.
Fig. 2 Same Steel as Fig. 1 but Containing 0.25 per cent Lead. Unetched. 1000 X.



THAT A SMALL LEAD CONTENT in steel improves its machinability is now well established. Evidence of this has been presented by Robbins (Robbins, F. J. "Character of Machine Performance of Lead-Bearing Steels," *Iron Age,* Vol. 142, Nov. 17, 1938, pp. 28-33), and much more is in the possession of those who are using such steels in everyday operation. Still further data will be presented in this article.

An improvement in machinability can be noted when less than 0.10 per cent lead is present, with increasing effect as the amount is increased up to almost 0.50 per cent. Slightly under that amount appears to be the limit for uniform dispersion.

Distribution of the Lead

The metallographer who is given a specimen of lead-bearing steel for examination, and who recalls the definite statements in metallurgical literature that "lead is insoluble in solid or liquid steel" may expect to see discrete pools of lead in the structure, just as he is accustomed to see them in leaded brass. He fails to find them and, indeed, from microscopic evidence only he is likely to doubt whether there is any lead present! Chemical analysis, however, tells him that lead is present. By employing sodium picrate plus sodium sulphide, the metallographer finally gets a stain that differentiates between lead-bearing and lead-free steels, but still fails to show up any discrete particles. As Figs, 1 and 2 show, the appearance of the inclusions in high sulphur screw stock is not altered.

That the inclusions in steels high in sulphur are modified by lead is indicated by the fact that a very small part of the lead present seems to be taken up by the sulphide inclusions and can be "sweated" out of the inclusions by heating to 900 deg. F. in hydrogen as Fig. 3 shows but the body of the steel shows no such sweat. Only a very small part of the lead present can be accounted for in this way.

Taking Sample of Steel from Open-Hearth Furnace for Testing Before Tapping (Courtesy: Inland Steel Co.)

Analyses of specimens taken from different locations in the ingot show a lack of the segregation one would expect to come about from settling of lead. This is, of course, true only when the lead is held below the limit of emulsification. Data to be cited herein will ordinarily refer to a lead content of 0.20 to 0.25 per cent. Experiments on "over-leaded" steels, in which agglomerations are present, show them to have slightly poorer physical properties than the steels in which the lead is uniformly dispersed.

There is evidence of a very slight enrichment of lead in the portions of the ingot latest to freeze or where advancing crystallization can trap unfrozen liquids. Fig. 4 shows, however, that nothing of the order of the trapping of large globules from a two liquid layer system has occurred in the case of lead properly added to steel. Indeed the lead is distributed more uniformly than carbon, manganese, or sulphur.

Is the Lead in Solution?

To evaluate the possibility that the books are wrong, i.e., that, contrary to all published data, lead can be held in true solid solution in ferrite, one next examines the steel for the usual effects of putting some other metal into solution, such as change in electrical conductivity, or of lattice dimensions. No detectable change in conductivity appears, and X-ray diffraction patterns, Fig. 5, show no change in lattice dimensions. Either the lead is not in solid solution in appreciable amount or it fails to show the expected influence. Critical point determinations show no clear effect.

From the positive evidence of very uniform dissemination and the negative evidence of lack of solid

ing, Lead-Bearing Steels

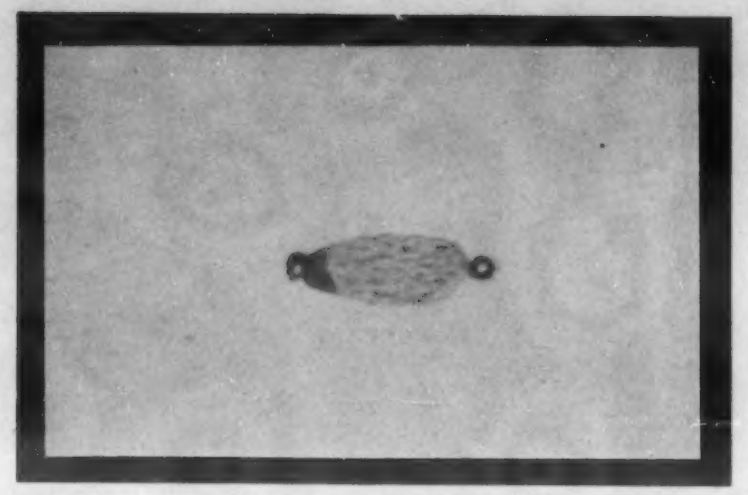


Fig. 3. Polished Sample of Leaded Screw Stock (0.25 per cent Lead) after Heating in Hydrogen to 900 deg. F. Lead globules at ends of sulphide inclusion. 1000 X.

solution, it appears probable that the lead is present in a submicroscopic state of subdivision, analogous to the dissemination of particles in a precipitation-hardened alloy or a grain size controlled steel, i.e., in a very different order of size and far more uniform dissemination than is the case in leaded brass.

Indirect evidence pointing in the same direction is the fact that emulsified lead does have a slight grain refining effect. Fig. 6 and 7 show samples of lead-free and lead-containing steel from the same heat, both normalized from 1600 deg. F. The steel with lead has a slightly, but appreciably, smaller grain size, and correspondingly, is very slightly less depth-hardening on quenching.

Mechanical Properties

Most of the readily machinable steels heretofore produced are given their easy machinability by some addition or method of processing which makes them more brittle, diminishing ductility and impact resistance. This necessary depreciation of mechanical properties limits the uses to which free-machining steels can be put. Hence any innovation producing enhanced machinability must be examined to see whether undue brittleness accompanies it.

The outstanding point about the lead-containing steels is that they are not brittle; instead, they are practically indistinguishable from similar lead-free steels. This was evident in the early experimental work on screw stock steels, as is shown in Table I.

A set of comparison commercial screw stocks was given impact tests over a range of temperatures. The compositions of the steels were:

	C	Mn	S	P	Si	Lead
B	0.18	0.73	0.111	0.021	0.015	None
A	0.18	0.81	0.134	0.022	0.015	0.13
C	0.18	0.74	0.191	0.022	0.015	0.25
D	0.18	0.75	0.127	0.021	0.014	0.26

Fig. 4. Lead Analyses on Split Ingots of Lead-Bearing Steel.

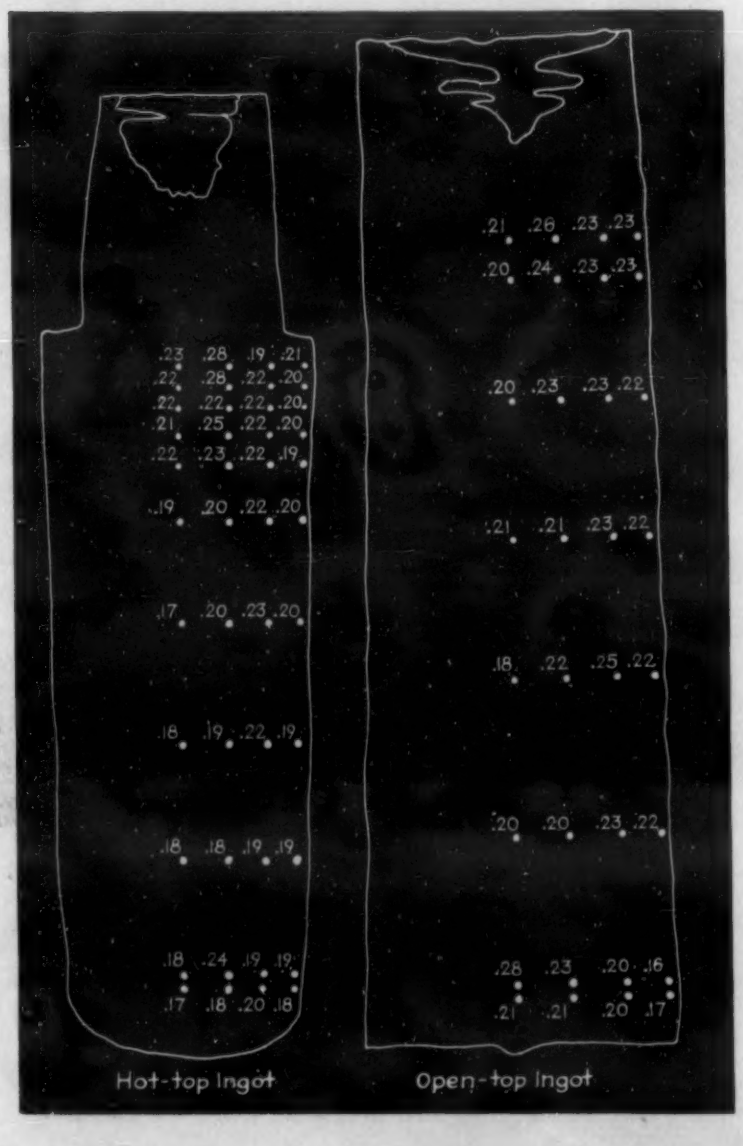


Table I—Composition and Mechanical Properties of Cold Drawn† Experimental Screw Stock Steels*

Heat No.	Composition, Per Cent						Yield strength, lbs. per sq. in.	Ultimate strength, lbs. per sq. in.	Per cent Elongation in 2 in.	Per cent Reduction in area	Charpy impact	Brinell Hardness	
	C	Mn	Si	P	S	Pb						Before cold drawing	After cold drawing
2962	.11	.63	.012	.017	.19	..	75,100	75,700	17.0	55.5	23.0	104	149
2966	.10	.55	.012	.019	.20	.12	72,000	73,500	18.5	56.5	20.0	105	141
2967	.11	.59	.010	.017	.21	.26	73,100	73,900	18.0	55.0	19.0	100	133
2968	.11	.58	.010	.019	.21	.34	71,100	72,500	19.0	56.0	19.5	99	135
2969	.11	.62	.008	.017	.23	.48	73,200	73,900	18.0	52.0	18.5	101	144
2965	.12	.57	.014	.018	.10	.09	70,700	70,800	20.0	58.0	24.5	95	134
2974	.14	.69	.248	.019	.20	..	76,800	78,400	16.5	52.5	19.5	115	156
2963	.11	.68	.238	.020	.21	.09	75,300	77,000	17.5	53.0	22.0	103	152
2973	.11	1.19	.228	.020	.19	..	81,000	79,700	17.0	54.5	22.0	118	151
2975	.12	1.20	.203	.020	.20	.09	79,000	79,000	19.0	56.0	31.5	113	143
2964	.15	1.11	.010	.020	.20	.09	75,600	75,100	16.5	55.5	23.0	111	133
2991	.10	0.29	.104	.016	.26	.39	71,500	75,800	17.5	50.5	17.0	105	143

* See Table XI for machinability indices. † Approximately 7% in one pass.

The specimens were normalized from 1600 deg. The data appear in Table II.

Table II—Results of Charpy Impact Tests on Leaded Screw Stock. Charpy (Key Hole) Impact Strength—Ft. Lbs.

Test Temp., Deg. F.		Steel B, No Lead	Steel A, 0.13% Pb	Steel C, 0.25% Pb	Steel D, 0.26% Pb
600		28.2	21.0	21.8	26.0
		32.0	24.0	19.2	28.0
	Av.	30.1	22.5	20.5	27.0
500		31.5	30.6	22.5	32.6
		33.1	29.0	23.5	29.0
	Av.	32.3	29.8	23.0	30.8
450		32.2	27.3	23.3	29.5
		31.2	27.1	23.5	28.7
	Av.	31.7	27.2	23.4	29.1
400		32.4	29.2	25.4	30.6
		33.0	27.8	27.4	29.4
	Av.	32.7	28.5	26.4	30.0
350		35.6	30.6	25.5	32.5
		33.0	31.4	26.5	31.9
	Av.	34.3	31.0	26.0	32.2
75		35.6	35.5	31.2	35.4
		37.4	34.1	28.6	34.4
	Av.	36.5	34.8	29.9	34.9
32		32.8	33.5	26.4	36.0
		33.8	33.1	26.6	34.2
	Av.	33.3	33.3	26.5	35.1
0		30.5	34.8	25.3	33.4
		28.9	32.8	26.3	34.0
	Av.	29.7	33.8	25.8	33.7
—25		28.0	30.4	25.4	28.6
		26.6	28.6	22.6	26.6
	Av.	27.3	29.5	24.0	27.6
—50		4.1(?)	...	21.6	28.1
		22.2	...	19.8	29.5
	Av.	...	...	20.7	28.8

The effect of cold drawing these steels was studied by drawing down to 0.448 in. dia. a tapered bar 0.450 in. dia. at one end, 0.505 in. at the other and making Izod impact tests.

The results are shown in Table III. Tests at 10, 20, 40, 80 and 160 days after drawing showed no appreciable change in impact properties. All the data are averaged in Table III.

Table III—Izod Impact in Ft. Lbs.

Per Cent Cold Reduction	B No Lead	A 0.13% Pb	C 0.25% Pb	D 0.26% Pb
3.1	67	66	55	68
6.5	56	54	45	58
10.1	49	46	39	50
13.4	45	45	37	45
16.5	42	41	34	43
19.5	40	39	33	41

From these data, it is indicated that the presence of lead in such steels does not alter their response to cold drawing.

Torsion tests were made on as-rolled "medium manganese" case hardening stock with and without lead. The results of the torsion tests are given below:

Specimen No.	Condition	Yield strength, lbs. per sq. in.	Ultimate strength, lbs. per sq. in.	Angle of twist in 10 in. gage length, deg.
J (regular)	As rolled	29,500	50,300	1840
J (regular)	As rolled	29,300	51,000	1875
JL (.18% Pb)	As rolled	29,500	54,500	2070
JL (.18% Pb)	As rolled	29,500	54,500	2070
2 (.21% Pb)	Cold drawn	52,800	53,500	2100
3 (.20% Pb)	Cold drawn	52,500	53,400	2030

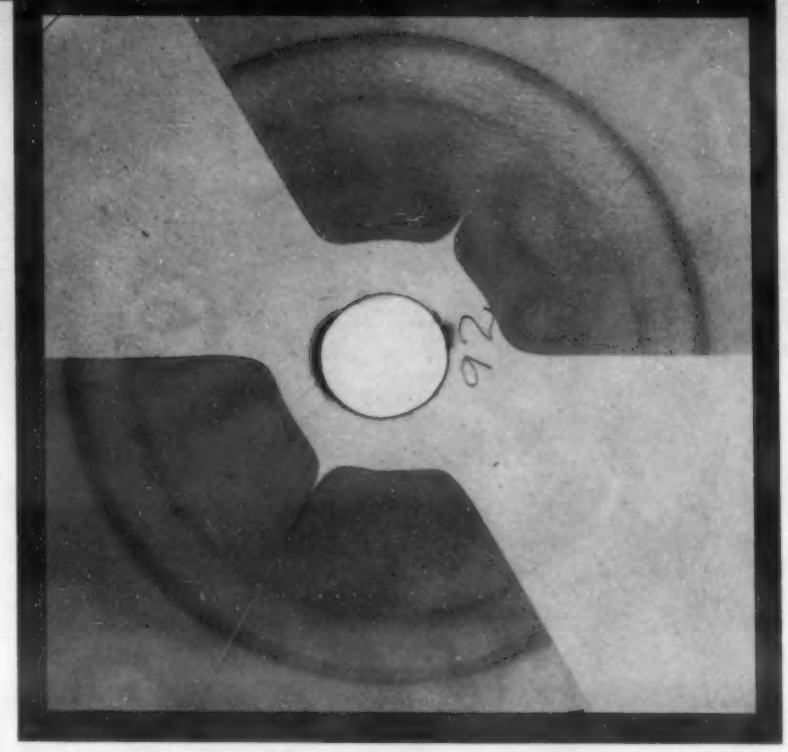
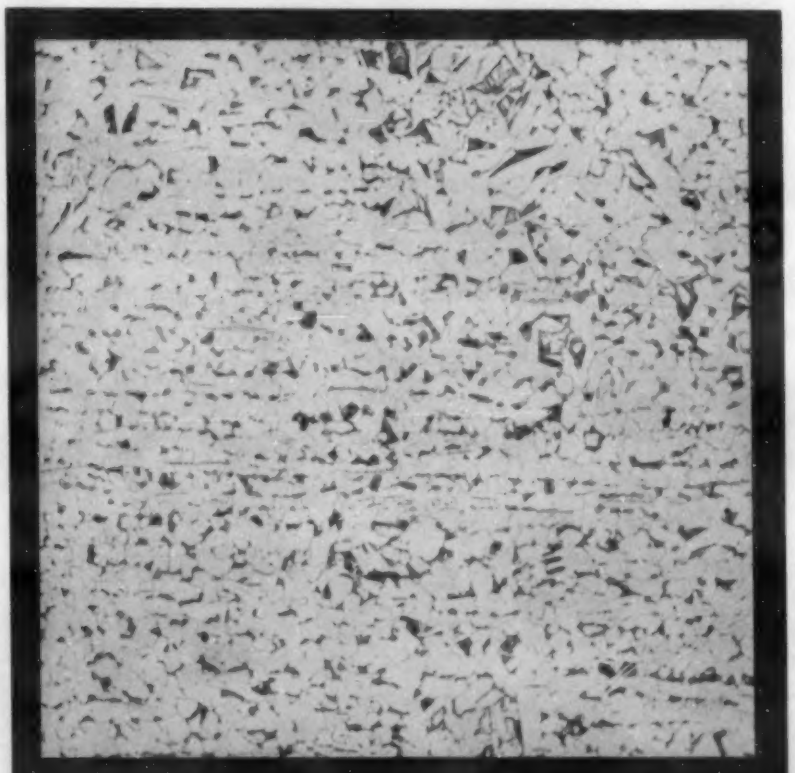
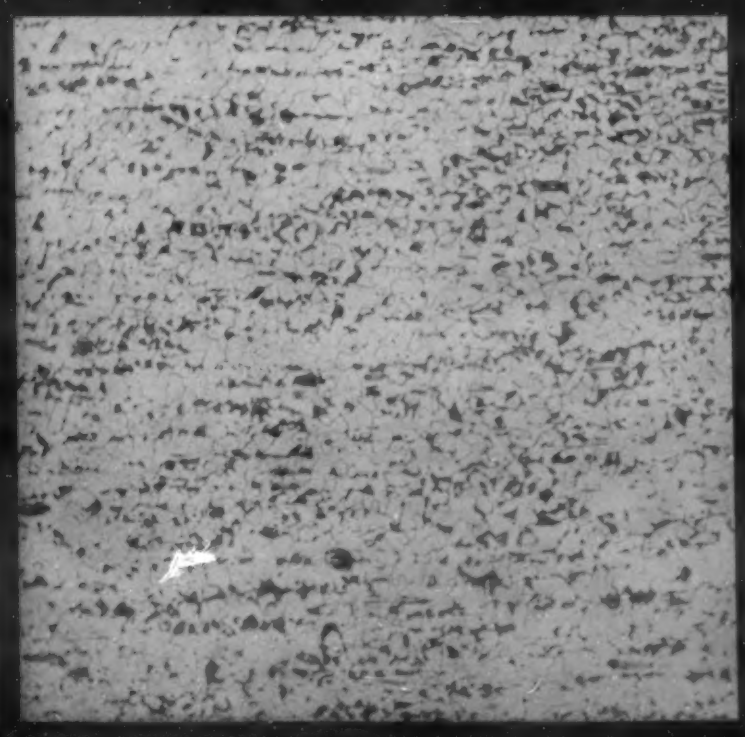


Fig. 5. Back Reflection X-Ray Diffraction Patterns of Leaded and Non-Leaded Steels. The dimensions of the iron lattice for both steels are identical.

Figs. 6 and 7. Both Steels from the Same Heat and Normalized from 1600 deg. F. Showing the Fine Grain Size of the Steel Containing Lead. Fig. 6 (Left) is a lead-free high sulphur, 0.15 per cent C steel, 100 X. Fig. 7 is a high sulphur, 0.15 per cent C steel containing 0.25 per cent lead.





Approximate analysis of this steel was:

C	Mn	P	S	Si
0.20	1.40	0.015	0.225	0.05

Commercial cold drawn SAE 1015 and 1115, with and without lead, give the following representative data:

	Tensile	Yield	Elong.	R.A.	Brinell	Charpy
1015 cold drawn 1 in. diam., no Pb	68,000	55,000	25.0	55.0	140	..
1015 cold drawn 1 in. diam., Pb *	67,000	55,000	27.0	56.0	137	..
1115 cold drawn ⅞ in. diam., no Pb	67,000	66,000	18.5	51.3	137	31
1115 cold drawn ⅞ in. diam., Pb *	71,000	69,000	18.5	51.3	143	27

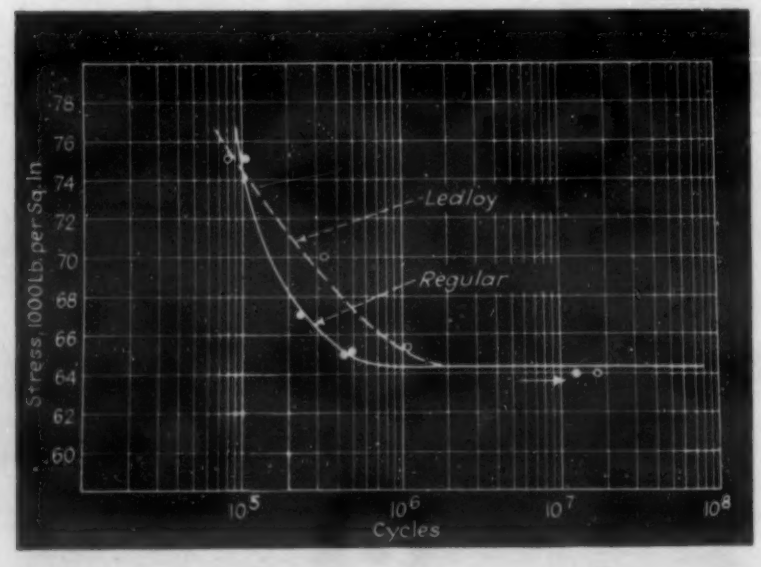
* 0.20 to 0.30% lead.

This lack of effect upon mechanical properties was intriguing, so the matter was followed up on stronger steels than screw stock. S.A.E. 1045 was made up into hot rolled 2 in. dia. bars with and without lead. The bars were quartered and 0.505 in. specimens made therefrom, being heat-treated in 0.01 in. oversize. Endurance specimens were heat treated in 0.5 in. dia. The specimens were drawn back to 255 Brinell and machined to size. The endurance specimens were 0.27 in. dia. The steels contained 0.46 C, 0.78 Mn, 0.025 P, 0.030 S, 0.18 Si, one being lead-free, the other containing 0.23 per cent lead. These 255-Brinell specimens showed:

	Tensile	Yield	Elong.	R.A.	Endurance Limit
1045 No Lead	121,500	86,000	19.0	54	64,000
1045 0.23% Pb	121,500	85,000	17.5	51	64,000

The endurance curves are shown in Fig. 8. The endurance limits are identical. That the practical identity of behavior of lead-containing and lead-free SAE-1045 extends over the usual range of draw temperatures for both water and oil quenched specimens is shown by Figs. 9 and 10.

Fig. 8. Endurance Curves for Regular and Lead-Bearing 1045 Water-Quenched and Drawn to 255 Brinell.



To evaluate the behavior of lead in alloy steels an experimental series of S.A.E. steels was made up as in Table IV, each being made with the amount of lead shown, and with no lead added.

Mechanical properties were determined on the low carbon members of this group in the normalized condition, and after cold drawing 7 per cent. The results are given in Tables V and VI.

The higher carbon steels of Table IV were tested as annealed, and oil quenched and drawn, with the results given in Tables VII and VIII.

(To be concluded)

Table IV—S.A.E. Steels Containing Lead

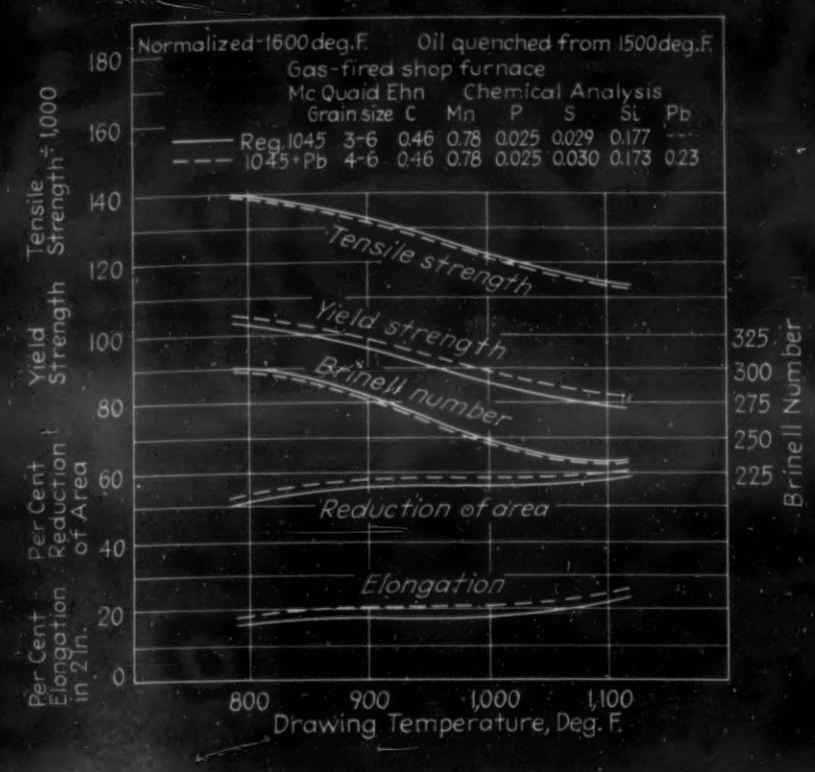
Type	C	Mn	Si	S	P	Ni	Cr	Mo	V	Cu	Pb
2315	0.17	0.65	.27	.027	.014	3.47	..	..	..	...	.17
2350	0.47	0.92	.30	.026	.013	3.46	..	..	..	...	.20
3115	0.16	0.59	.25	.024	.011	1.25	.65	..	..	...	.14
3145	0.45	0.83	.29	.027	.012	1.25	.66	..	..	...	.18
4130	0.30	0.75	.19	.028	.012	...	.66	.20	..	...	.12
4320	0.20	0.53	.14	.023	.013	1.75	.65	.33	..	...	.11
4340	0.41	0.75	.22	.022	.012	1.74	.65	.34	..	...	.17
4615	0.16	0.62	.21	.023	.012	1.76	..	.25	..	...	.16
4640	0.40	0.75	.21	.025	.012	1.78	..	.24	..	...	.18
5120	0.20	0.57	.21	.024	.010	...	.75	..	..	...	.12
5150	0.52	0.86	.22	.030	.012	...	.94	..	..	...	.17
6115	0.15	0.58	.21	.026	.010	...	.91	..	.18	...	.16
6150	0.52	0.84	.22	.031	.010	...	.93	..	.18	...	.21
T1330	0.31	1.77	.23	.031	.015	...	..	..	..	...	.15
Hi-Steel	0.20	0.54	.10	.029	.120	...	..	..	..	1.00	.18

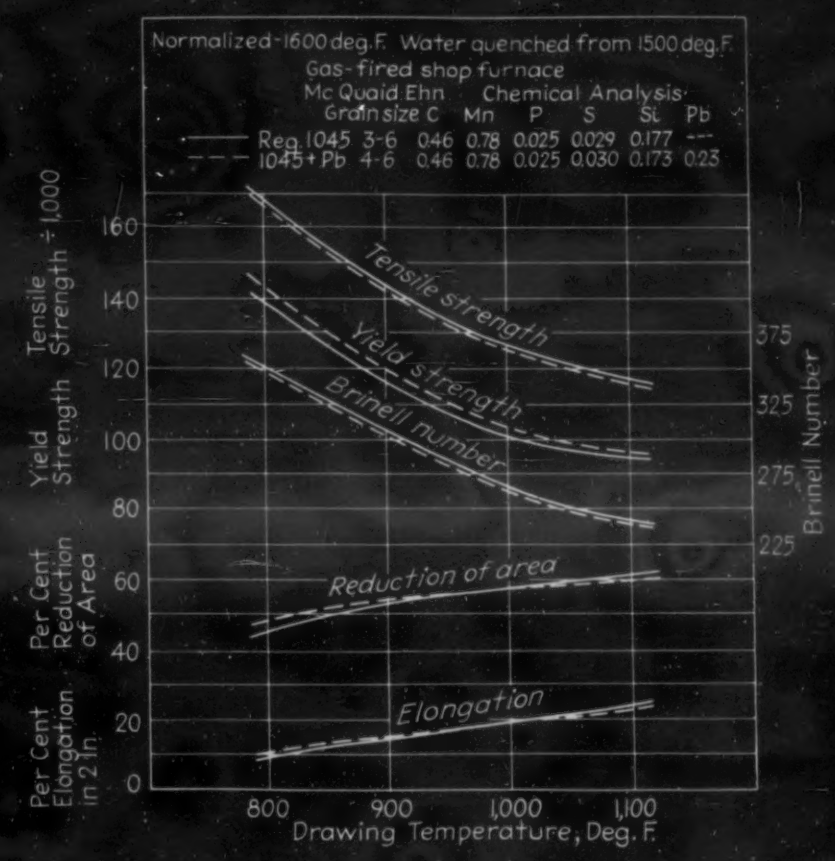
Table V—Low Carbon S.A.E. Steels Normalized in ¾ In. Dia.

Type	Pb	Norm. Temp., deg. F.	Tensile	Yield	Elong.	R.A.	Brinell	Charpy
2315	...	1500	86,500	58,000	31	63	163	44
2315	0.17	1500	88,000	62,500	31.5	61	163	39
3115	...	1600	79,500	57,000	33	70	146	55.5
3115	0.14	1600	79,000	60,500	34	67.5	149	49
4320	...	1600	115,000	80,000	21	57	207	35
4320	0.11	1600	115,000	77,500	21	65.5	207	33
5120	...	1700	76,000	56,000	35	68	131	48
5120	0.12	1700	75,500	54,000	35	66	137	51
6115	...	1700	75,000	59,500	35	73	131	58.5
6115	0.16	1700	75,000	58,000	35.5	70.5	126	53
Hi-Steel	...	1700	84,000	64,000	31	57	152	24
Hi-Steel	0.18	1700	84,000	64,500	30	56	156	22.5

Table VI—Low Carbon S.A.E. Steels Cold Drawn

Type	Lead	Tensile	Elong.	R.A.	Brinell	Charpy, Ft.Lb.
2315	...	104,000	15	56	187	35
2315	0.17	103,000	14	53.5	189	33.5
3115	...	89,000	20	67	170	48.5
3115	0.14	88,000	20.5	64	169	45
4320	...	126,000	12.5	53	226	27.5
4320	0.12	125,000	12.5	52.5	229	28.5
4615	...	106,000	14.5	58	185	40
4615	0.18	108,000	14	56.5	189	36
6115	...	82,000	23.5	70	149	55
6115	0.16	83,000	22.5	68	156	47.5
Hi-Steel	...	102,000	16.5	48	183	17.5
Hi-Steel	0.18	105,000	15	54	189	11.5


Normalized-1600 deg.F. Oil quenched from 1500 deg.F.
Gas-fired shop furnace
Mc Quaid Ehn Chemical Analysis
Grain size C Mn P S Si Pb
Reg. 1045 3-6 0.46 0.78 0.025 0.029 0.177 ---
1045+Pb 4-6 0.46 0.78 0.025 0.030 0.173 0.23
Tensile strength
Yield strength
Brinell number
Reduction of area
Elongation
Tensile Strength ÷ 1,000
Yield Strength
Per Cent Reduction of Area
Per Cent Elongation in 2 In.
Brinell Number
Drawing Temperature, Deg. F.


Normalized-1600 deg.F. Water quenched from 1500 deg.F.
Gas-fired shop furnace
Mc Quaid Ehn Chemical Analysis
Grain size C Mn P S Si Pb
Reg. 1045 3-6 0.46 0.78 0.025 0.029 0.177 ---
1045+Pb 4-6 0.46 0.78 0.025 0.030 0.173 0.23
Tensile strength
Yield strength
Brinell number
Reduction of area
Elongation
Tensile Strength ÷ 1,000
Yield Strength
Per Cent Reduction of Area
Per Cent Elongation in 2 In.
Brinell Number
Drawing Temperature, Deg. F.

Figs. 9 and 10. Physical Property Charts of Oil (Left) and Water-Quenched SAE 1045 Steel with and without Lead.

Table VII—Higher Carbon Steels of Table IV—Annealed

Type	Pb	Annealed at	Tensile	Yield	Elong.	R.A.	Brinell	Charpy, Ft.Lb.
2345	...	1475	117,000	71,500	22.5	46	217	14
2345	0.20	1475	118,000	74,500	23	47.5	217	14
3145	...	1475	104,500	60,000	24.5	49	192	18
3145	0.18	1475	104,500	59,000	24.5	49	187	18.5
4130	...	1600	85,000	54,000	28	50.5	146	25.5
4130	0.12	1600	86,000	55,000	27.5	50	143	27
4340	...	1525	138,000	134,500	17	47.5	241	13.5
4340	0.17	1525	142,000	133,500	17.5	46.5	241	14
4640	...	1525	101,500	62,500	21.5	41	187	12.5
4640	0.18	1525	101,500	62,000	22	40.5	187	13
6150	...	1530	105,500	61,500	24.5	49	197	15
6150	0.21	1530	101,500	61,000	25	49	192	16.5
T1330	...	1525	95,000	64,500	28	56	156	16.5
T1330	0.15	1525	95,500	66,500	28	56	159	17
5150	...	1550	105,000	45,500	21.5	36	179	10
5150	0.17	1550	106,500	46,000	21.5	35.5	179	10

Table VIII—Higher Carbon Steels of Table IV—Oil Quenched and Drawn

Type	Lead	Quenched from	Drawn	Tensile	Yield	Elong.	R.A.	Brinell	Charpy
2345	...	1475	1150	125,500	85,500	20	42.5	220	23.5
2345	0.20	1475	1150	124,000	94,000	20	45.5	223	23
3145	...	1475	1200	156,000	143,000	19.5	58	290	29.5
3145	0.18	1475	1200	151,000	140,000	19	58.5	290	30
4130	...	1600	1125	133,000	117,000	20	65	245 *	45
4130	0.12	1600	1125	119,000	99,000	20.5	61.5	217 *	44
4640	...	1525	1200	147,500	133,000	20	61	270 *	35
4640	0.18	1525	1200	140,000	120,000	20	58	260 *	35
6150	...	1550	1200	164,500	154,500	18	55.5	315	24.5
6150	0.21	1550	1200	163,500	148,000	16.5	53	315	23.5
T1330	...	1525 †	1125	122,500	113,000	19	58	230	21.5
T1330	0.15	1525 †	1125	121,000	113,000	19.5	55	225	20
5150	...	1550	1200	152,500	131,000	19	57.5	280	30
5150	0.17	1550	1200	150,000	127,000	18	57	280	28

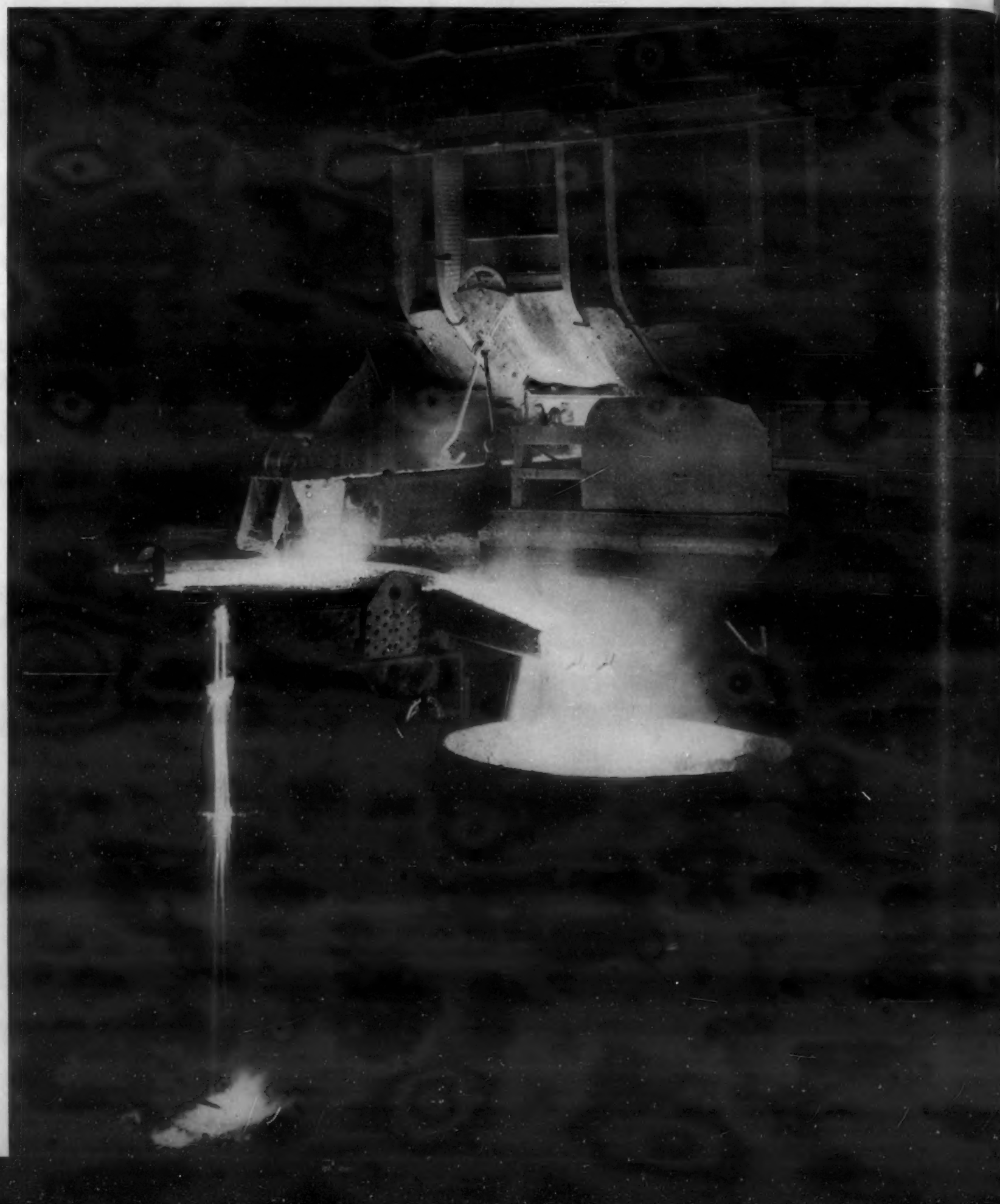
* Variations probably caused by accidental differences in adhesion of scale in quenching. † Water quenched.

Grain Size of Steel

—A CRITICAL REVIEW by N. F. WARD and J. E. DORN

Associate Professor of Mechanical Engineering and Assistant Professor of Mechanical Engineering, Respectively, University of California, Berkeley, Calif. Prof. Dorn was formerly Research Associate at Battelle Memorial Institute, Columbus, Ohio.

Tapping a 160-Ton Heat of Open-Hearth Steel *(Courtesy: Gary Works, Carnegie-Illinois Steel Co.)*



Nowadays "grain size" specifications are as important a factor in the purchasing of steel as are the traditional composition requirements familiar to everyone. The informed consumer realizes the great extent to which the properties in which he is interested are affected by austenitic grain size, and the progressive producer applies suitable grain size control methods in his steel-making operations. All industrial factors have a common interest in methods of testing and reporting grain size.

This unusually comprehensive correlation gives in simple, workable form all the information the metallurgical engineer needs to know about austenitic grain size—what it is (and what it is not*); how to determine it, in shop or in laboratory; how it influences the engineering properties of steel; and how it may be controlled in steelmaking operations. The information has been carefully brought up-to-date, so that this series of articles should be one of the most useful, complete expositions of this important subject currently available.—The Editors.*

THE INTEREST THAT HAS BEEN AROUSED by the investigations on austenitic grain size and its resultant effects on the properties of commercial steels may be directly attributed to the application of mass production methods of fabrication. For the successful application of such methods, many heats of steel must have the same physical and mechanical properties, within rather narrow limits, after identical mechanical and heat treating operations. If, for example, different heats of the same chemical analyses exhibit different responses to heat treatment, many articles will fail to pass inspection. In order to alleviate this wasteful condition, it may be possible to specify a new successful heat treatment for each new heat of steel. When this is done, however, mass production is no longer in effect, and its benefits are lost. The success or failure of mass production methods, therefore, depends not only on uniformity of chemical analysis of steels but also on the uniform response of many heats of a specified steel to thermal and mechanical treatment.

For many years it was known that a specified chemical analysis alone was not a guarantee of uniformity, which was so necessary in modern methods of fabrication. Often heats of steels that had the same reported chemical analyses varied greatly in their responses to heat treatment. For purposes of discussion, various heats of the same steel were personified and their inherent differences were attributed to differences in "character," "personality," "quality," "timbre," or "body." As Benedicks and Löfquist[67] have shown, at least two properties were embodied in these terms; they were (1) sufficiently good mechanical properties, and (2) a broad heat treating range. Obviously such terms were decidedly inappropriate and indefinite.

It was recognized that the differences in response to heat treatment were dependent upon the deoxidation practice involved in making the heat. If the heat were killed with strong deoxidizers such as aluminum, titanium, etc. one set of properties was obtained, and if killed only with silicon, or if unkilled, other markedly different properties resulted.

Unfortunately, additions of one or two pounds per ton of a strong deoxidizer were frequently used to produce sound ingots in spite of poorly controlled deoxidation, and this often led to dirty steels, or steels with abnormally low ductility. Many of the more conservative steel producers, for this reason, denounced this procedure and referred to the method as "doping" or "dosing" a heat. At present, such accusations have no merit. Aluminum and other deoxidizers frequently are added to perfectly sound heats to introduce certain metallurgical and physical effects which are highly desirable for specific properties in the finished product. Each application, however, demands special attention, and although thoroughly deoxidized steels may be better for some purposes, for others, steels deoxidized with silicon only are more satisfactory.

No introduction to the subject of austenitic grain size is complete without paying tribute to McQuaid and Ehn[4] and Ehn[2, 3] for their pioneering in this field. Their investigations on the "Effect of Quality of Steel on Case Carburizing Results" stimulated an overwhelming number of research projects which eventually led to a new chapter in the history of heat treating. Industry and science alike will remain indebted to them for this truly American contribution to ferrous metallurgy.

Although much thought and research has been devoted to the study of austenitic grain size, and proper control within reasonable limits is a commercial fact, many aspects of the subject are still under discussion. This is especially true concerning the theories proposed for the various phenomena observed in investigations on austenitic grain size. Although the purpose of this paper is to present a correlated abstract of the most important work on this subject, it was thought advisable to consider, critically, some of the theories that have been proposed, in an effort to evaluate their relative merits. In so doing, the authors are fully aware that they are offering their prejudiced opinions on the subject. This, however, should not be too disconcerting to the reader, since a definite attempt to differentiate between the factual knowledge and the various theories will be made.

Methods of Revealing the Austenitic Grain Size

The phenomena that are correlated with the austenitic grain size in this review refer only to steels that may be hardened by quenching. The typical

austenitic steels that form stable austenitic phases at atmospheric temperatures, and all of the alloy steels that retain large quantities of untransformed austenite on slow cooling to atmospheric temperatures, do not necessarily exhibit the peculiarities generally attributed to austenitic grain size. In fact, the austenitic grain size is highly important only in plain carbon and low alloy steels. With increasing alloying content, this significance diminishes because of the paralleling influence of alloying and fine austenitic grain size on toughness.

When hardenable steels are heated above the upper critical temperature, a transformation takes place, and a single phase, austenite, is formed. It consists of a solid solution of carbon and the various other alloying elements in the face-centered cubic lattice of gamma iron. The entire piece of steel, however, generally contains a great number of such identical lattices variously oriented with respect to one another, and cemented together by a series of atoms which, apparently, gradually change their positions from sites consistent with the orientation of one lattice to those consistent with the orientation of an adjacent lattice. Each lattice is an austenitic grain which is separated from adjacent austenitic grains by more or less amorphous regions called the austenitic grain boundaries. It is the size of these austenitic grains that influences so profoundly the properties of steels.

If a steel is cooled below the lower critical temperature (eutectoid temperature), the austenitic grains decompose, forming an entirely new set of grains. Slowly-cooled steels form new grains of ferrite and carbide while steels cooled more rapidly than the critical quenching rate yield new martensitic grains. Among other factors, the size of the new grains depends upon the cooling rate. It is not possible, therefore, to determine the austenitic grain size of steels at atmospheric temperatures by microscopic examination of polished and etched specimens, unless special precautions are taken to reveal the size of the former austenitic grains after they have decomposed. (See Fig. 1.)

All methods of revealing the size of the former austenitic grains depend upon the fact that grain boundary material, in general, exhibits greater chemical and physical activity than do the grains themselves. The greater activity of the grain boundary material not only reveals a record of the decomposed austenitic grain size, but also accounts for the major differences in "quality" between fine and coarse grained steels.

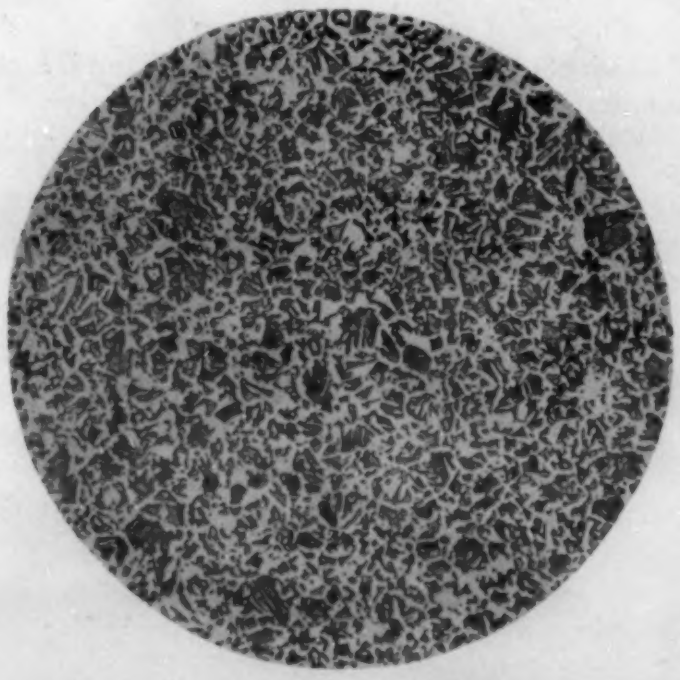


Fig. 1. Annealed S. A. E. 1035 Steel. The size of the pearlitic patches does not reveal the size of the former austenitic grains.

By Carburizing

Inasmuch as the first investigations on austenitic grain size were the results of the efforts of McQuaid and Ehn[4] to determine the cause of differences in the "quality" among steels of the carburizing type, it is not surprising that carburization was the first method employed to reveal the austenitic grain size. The method consists of heating a small sample of the steel packed in a pot with carburizing compound or exposed to carburizing gases for any desirable length of time, at any suitable temperature in the austenitic range. The time and temperatures are limited to those that will produce a visible amount of hypereutectoid case. The sample may be cooled with the furnace, or removed from the pot or furnace and cooled in the air. During cooling through the critical range, precipitation of proeutectoid carbides will occur at the austenitic grain boundaries because of the greater activity at these points. The specimen is then sec-

Fig. 1a. A Coarse-Grained Steel as Revealed by Carburizing.

(Courtesy of O. E. Harder)

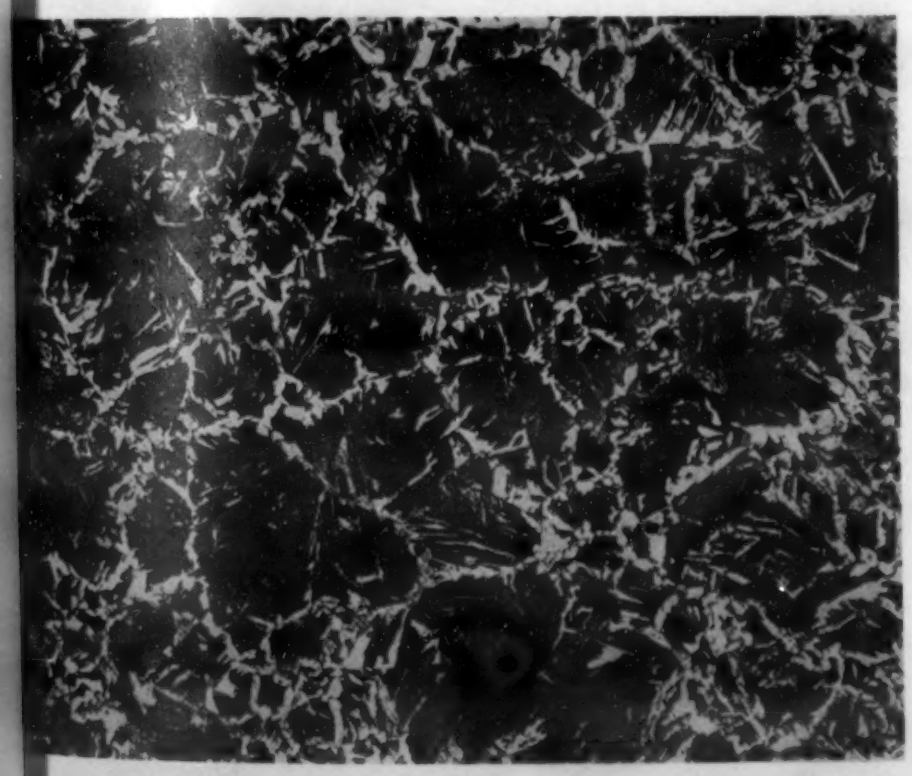
tioned, polished, and etched for a few seconds in nital or picral. Microscopic examination will reveal a network of carbides about islands of pearlite. Each pearlitic island represents a single former austenitic grain. (See Fig. 1a.)

If the conditions are such that abnormality results (coalescence of the carbides and divorcement of ferrite) it may be difficult to estimate the austenitic grain size. Rapid cooling of the sample after carburizing reduces this tendency[10], but if the austenitic grain size is very small, unequivocable outlining of the austenitic grains may be difficult. Other peculiarities may be noted, among them the presence of a fine grained outer rim of and a coarse grained inner zone of the case.[23] This has been attributed to the reaction of oxygen from the carburizing gases on the surface, and its diffusion inward during carburizing. According to this explanation, oxygen diffuses less rapidly than carbon, and therefore the fine grained rim occurs only on the outer zone of the case. Less frequently the reverse is noted.[32] In such instances the case is coarse and the core remains fine. Although no satisfactory answer to this anomaly has been offered, two explanations are possible. The coarsening of the case may be due to the greater displacement of the higher carbon regions from their upper critical temperatures or, more likely, the reduction of the growth inhibiting substances by the carburizing gases or carbon.

Carburizing is undoubtedly the most accurate method for revealing the austenitic grain size of the carburizing grades of steels. It is necessary, however, to reproduce all conditions such as temperature, time, and gas composition of the actual carburizing cycle in order to correlate the grain size as revealed by the test with the physical properties actually obtained in practice, for, as will be shown later, these factors influence the grain growth.

Indiscriminate use of the carburizing method of determining the austenitic grain size for correlation with heat treatments other than carburizing often yields incorrect results. This has been the thesis of a number of recent papers on austenitic grain size. Vilella and Bain[79] and Herty, McBride and Hollenback,[85] have stressed the fact that temperature, time, composition of gases, and carbon content of the steel influence its austenitic grain size. It is essential, therefore, to perform the grain size determination under conditions that accurately reproduce the actual heat treatment, at least up to the cooling or quenching step. Since little or no change in the austenitic grain size takes place during cooling, the test sample may be removed from the furnace at this time and treated to reveal the austenitic grain size by any of the following methods that is suitable for the steel that is being investigated.

Fig. 2. Hypoeutectoid Steel Cooled in Air. The size of the previous austenitic grains is revealed by the proeutectoid ferrite about the pearlitic patches.



By Slowly Cooling Hypereutectoid and Slightly Hypoeutectoid Steels

The slight excess of carbide in hypereutectoid steels and the slight excess of ferrite in hypoeutectoid

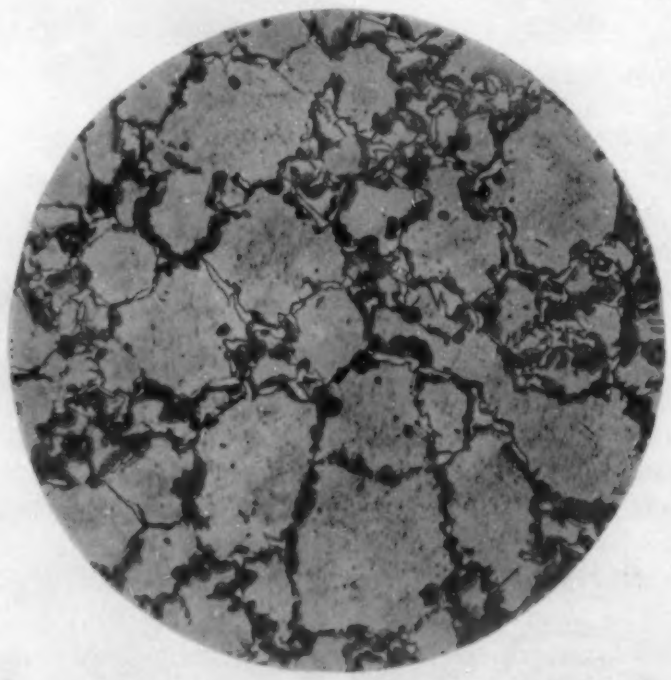


Fig. 3. Arrested Quenched S. A. E. 1040 Steel. During the arrest free ferrite formed at the austenitic grain boundaries. The grain boundaries are also outlined with pearlitic nuclei, illustrating that the decomposition of austenite begins at these points of greatest activity. The core of each austenitic grain was transformed to martensite by the final quench.

steels permits the use of a simple method for outlining the austenitic grains of these steels. When hyper- or hypoeutectoid steels are air-cooled or cooled with the furnace, proeutectoid carbide or proeutectoid ferrite, respectively, precipitates in the boundaries of the decomposing austenitic grains.[35] Each pearlite patch in the network of carbide or ferrite represents a former austenitic grain. This is illustrated in Fig. 2.

It is a simple matter to correlate the austenitic grain size with the properties of steels after any given heat treatment by this method. The sample is sawed, polished, etched with nital or picral, and investigated under the microscope. For carbon concentrations of 0.45 per cent C to 0.55 per cent C and for 1.10 per cent C and above, air cooling is satisfactory, but for concentrations of 0.55 per cent C to 0.70 per cent C and 1.00 per cent C to 1.10 per cent C, cooling with the furnace must be used in order to permit a sufficiently visible precipitation of carbide or ferrite in the boundaries of the austenitic grains. Inasmuch as both grain size and alloying influence the rate of carbide or ferrite formation during cooling, the above recommendations are only approximately valid for all steels. Since the steels are examined at the center of the sawed sections, the results of this and most of the following methods are not influenced by furnace gases, scaling, or other surface peculiarities.

By Arrested Quench

For steels of lower carbon content, the size of the pearlite patches is less indicative of the former austenitic grain size in slowly cooled samples. Special methods, therefore, were devised for revealing the austenitic grain boundaries of low carbon steels. The arrested quench method of Grossman[32] was one of the first reported. After the heating cycle has been carried out, up to the significant cooling operation, a small sample of the steel is removed from the furnace, placed in a bath of molten lead or salt at some temperature between the upper and lower critical temperatures for that steel, held at this temperature for a few seconds, and then quenched in water. During the short time at the arrest, proeutectoid ferrite forms at the boundaries of the decomposing austenitic grains. The subsequent quench transforms the remainder of the austenitic grain to martensite. A polished section of the specimen, etched in nital or picral shows a network of ferrite or, in some instances, fine pearlite, enveloping martensitic areas. Each martensitic area represents a single decomposed austenitic grain. Fig. 3 shows the microstructure obtained by this treatment.

The time and temperature of the arrest for the best results cannot be readily predicted.[69] Several attempts are generally required to secure only a narrow ferrite network about the former austenitic grains since the rate of ferrite formation is dependent not only on the carbon concentration but also on the actual austenitic grain size and alloying of the steel. If the proper adjustment is made, however, this method is very delicate for extremely fine austenitic grains which very often fail to respond properly to other methods. The gradient quench has superceded the arrested quench to a large degree.

By Gradient Quench

Vilella and Bain[79] have discussed various methods of revealing the austenitic grain size and demonstrated that the gradient quench is undoubtedly the most adaptable method yet reported. It may be applied over the entire range of carbon concentration and seldom fails to give excellent boundary markings. According to their recommendation, one end of a sample of steel about 1½ in. long and ¼ in. in diameter should be quenched vertically in ½ in. of brine following the pertinent heat treating operation. The sample is then sectioned longitudinally and polished and etched, nital or picral being satisfactory. When investigated under the microscope, a series of structures are revealed which range from light-etching martensite at the end that was immersed in the brine to the dark-etching carbide and ferrite at the end that was exposed to the air. Proceeding from the end that has a martensitic structure and investigating the changes in appearance as the less rapidly cooled sections are brought to view, a zone will be reached wherein the martensitic areas are outlined with fine pearlite. The network of dark-etching fine pearlite about the light-etching martensitic patches represents the former austenitic grain boundaries. If the specimen is examined nearer the air-cooled end, a second useful zone will be found. In this region a network of ferrite about areas of fine pearlite delineates the boundaries of the former austenitic grains. These two zones are shown in Figs. 4 and 5, respectively.

For Eutectoid Steels

It is slightly more difficult to reveal the austenitic grain size of eutectoid steels than steels of either hypo- or hypereutectoid compositions. It is possible, occasionally, to obtain sufficiently contrasting markings by quenching a 1-in. diameter section in brine. For most steels treated in this manner there will be a center zone which has a fine pearlitic network about martensitic areas, each of which represents a former austenitic grain.

The arrested quench may also be used for eutectoid steels. Bain and Vilella[65] have shown, however, that the gradient quench may be used even in the case of eutectoid steels. In such steels only one zone, namely that consisting of a fine pearlitic network about martensitic areas will be found. Because of its universal adaptability, this method is recommended.

By Etching Martensite

It is often desirable to learn what the austenitic grain size was, previous to quenching, when the only sample available has a martensitic structure. This may be accomplished by etching a polished section of the sample for 5 to 15 sec. in an etchant consisting of 5 cc. of concentrated hydrochloric acid, 95 cc. of alcohol, and one grain of picric acid, and results in a structure similar to that in Fig. 6. This method has been accredited to Archer by some authors[35, 77] and to Miss Baeyertz by others.[32] It is often difficult to develop the structural differences by this method and it has been stated that slight tempering enhances the differences. Snyder and Graff[99] have developed an etchant for tool steels in the martensitic state; Fig. 7 shows the grain size delineation obtained on hardened and on tempered tool steels.

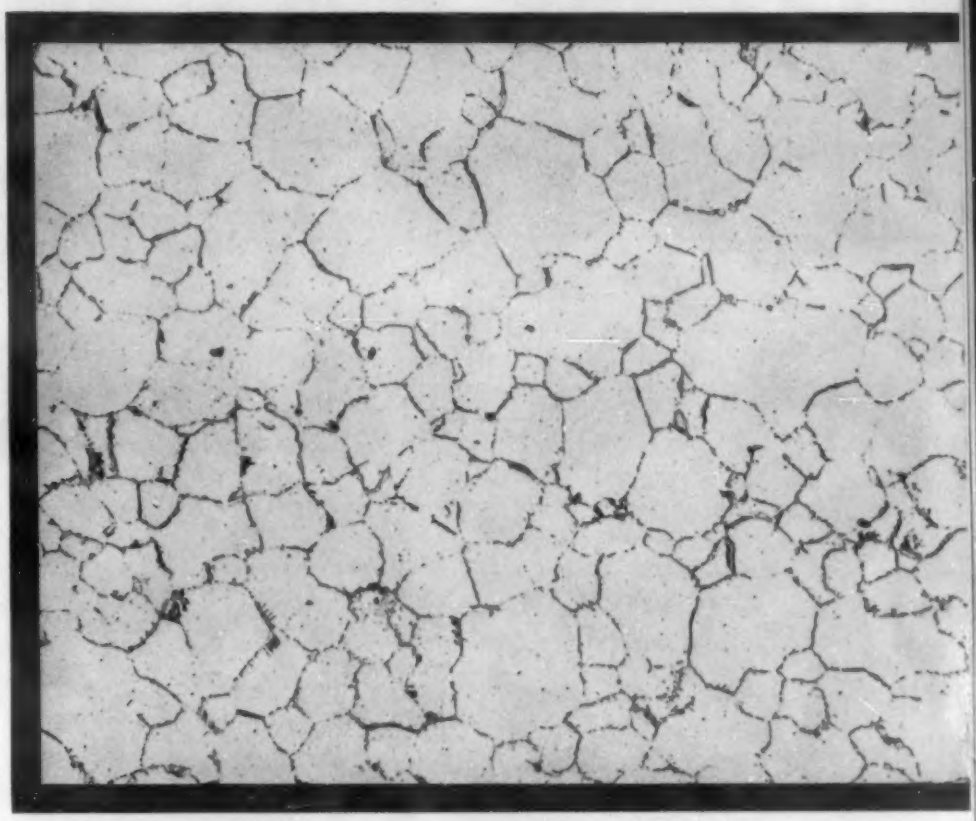
By Preferential Oxidation

It has been reported that suitable marking of the austenitic grain boundaries was obtained by heating a polished specimen in an electric furnace and quenching.[71, 89] The oxide scale was removed and the specimen was repolished and etched with nital. The former austenite grains were marked by accumulated oxides and envelopes of ferrite resulting from decarburization. This method, however, is not recommended since oxides or oxygen are often efficacious in producing finer grains.

By Preferential Vaporization

Preferential vaporization of the grain boundary material has been reported as a method of delineating the austenitic grain boundaries of low carbon steels and carbonless gamma phase alloys as well as pure iron. (See Hemmingway and Ensininger, *Trans. Amer. Inst. Mining Met. Engrs.*, Vol. 67, 1922, p. 392.) A highly polished specimen is heated in a vacuum. After cooling, microscopic examination reveals a network of deep furrows. Although it is generally agreed that these furrows represent the former

Fig. 4. Gradient Quenched S. A. E. 1045 Steel. This photomicrograph illustrates the first zone wherein the previous austenitic grain size is revealed by fine pearlite about martensitic cores.



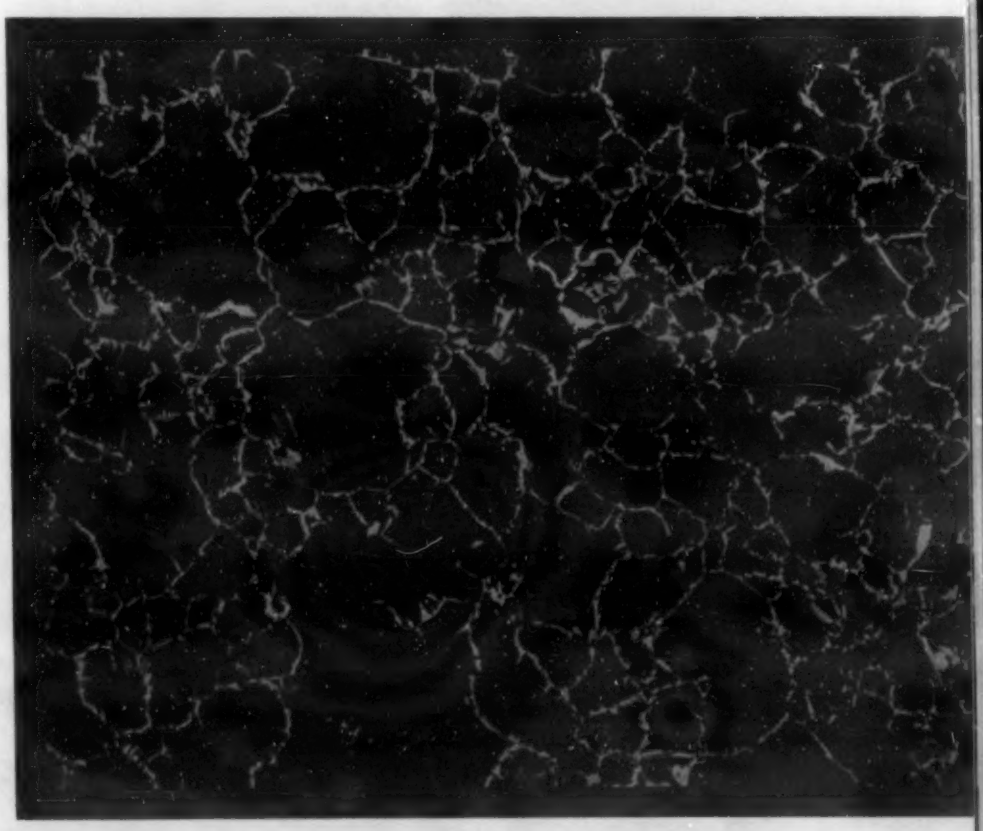


Fig. 5. Gradient Quenched S. A. E. 1045 Steel. This photomicrograph illustrates the second zone wherein the previous austenitic grain size is revealed by proeutectoid ferrite about pearlitic patches.

Fig. 6. (Left) Austenitic Grain Size as Revealed by a Special Martensitic Etch. (Courtesy of J. R. Vilella)

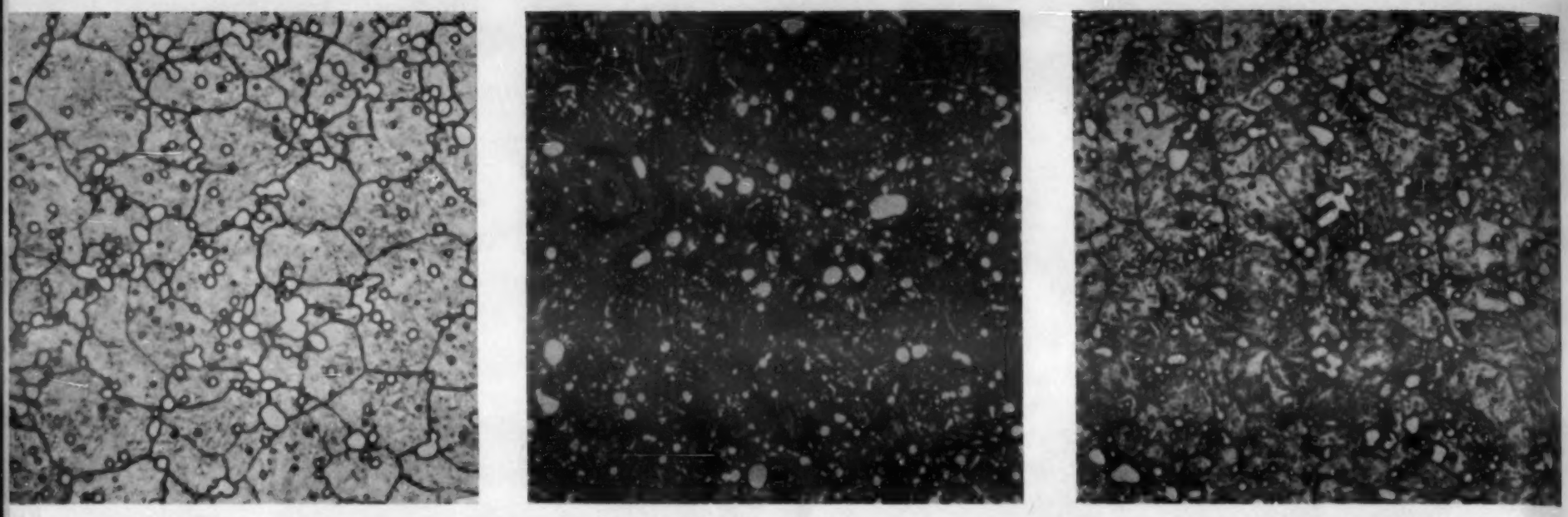


Fig. 7. Photomicrographs (x1000) Illustrating the Application of a Special Etchant for Tempered High Speed Tool Steels According to Snyder and Graff.[99] The photomicrograph at the left shows the size of the previous austenitic grains of a tool steel in the hardened condition etched with the usual 4 per cent nital solution. The same etchant is unsatisfactory for tempered tool steels as shown in the middle picture. The use of a special etchant gave the result shown at the right on a tempered tool steel. The boundaries of the former austenitic grains are readily discernible. The grain size was reported as No. 20 according to the scale developed by Snyder and Graff. (Courtesy of R. W. Snyder)

austenitic grain boundaries, there is some question as to whether they originate from preferential vaporization or processes that occur on cooling. Although this method was one of the first employed, it is rather limited to purely scientific investigations. For this reason it is of little interest in the present work on austenitic grain size.

The McQuaid-Ehn Test

Although the McQuaid-Ehn test[4] for austenitic grain size is, at present, of historical interest only, it has frequently been used in the past, and therefore deserves critical attention. The test is a special case of the carburizing method previously given. It consists of carburizing the specimen for 8 hrs. at 1700 deg. F. The results of a single test are then extrapolated to any other temperature, any duration of time at temperature, and to any furnace atmosphere. The possible influence of the added carbon is neglected. That is, it was assumed that this single test would differentiate between "inherently" coarse grained steels and "inherently" fine grained steels. Unfortunately, as will be demonstrated in the section on the formation of austenitic grains and their growth characteristics, steels do not exhibit "inherent" grain sizes.[76] In fact, recent experiments on the effect of pre-austenitic structures and pre-heat treating operations on the growth characteristics of austenitic grains indicate that even the expression "inherent characteristics of austenitic grain growth" is untenable. It is highly important that the word "inherent" be omitted from metallurgical discussions on austenitic grain size since inherent denotes unalterable properties.

Herty, McBride, and Hollenback,[85] as well as Vilella and Bain[79] and others have demonstrated that the application of the principles involved in the McQuaid-Ehn test very frequently lead to incorrect conclusions. It is now generally agreed that any test of the austenitic grain size should reproduce all of the steps in the actual heat treating cycle to be employed, up to the cooling or quenching operation. At this instant any suitable method of cooling to reveal the austenitic grain size may be used. An indisputable correlation between the physical properties and the austenitic grain size can be assembled for any recommended or suggested heat treatment if these principles are followed. *(To be continued)*

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PREPARATION OF

Rotating Beam Fatigue Specimens

by Q. HENDERSON and W. B. SEENS

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Our editorial director, when visiting the research laboratory of the Steel Corporation, was so much impressed with the outfit he saw there for longitudinal polishing of fatigue specimens that he asked that a description of the procedure be written for METALS AND ALLOYS. *This article is the result of that request.*

It emphasizes the necessity for the absence of circumferential scratches or tool marks and the need of perfect concentricity in order to avoid vibrational stresses and secure uniform and reliable results from a series of tests. The procedure for forming and finishing specimens to be tested in the R. R. Moore rotating beam fatigue machine is described.—The Editors.

IN THE ROTATING BEAM fatigue test the fiber stress is a maximum at the surface of the reduced section of the specimen; consequently the condition of this surface is a significant factor in the results of such tests. Essential identity of the surface is especially necessary in a group of like specimens to be tested at a series of loads as a means of determining the endurance limit of a metal; otherwise there may be a considerable degree of uncertainty in the result derived from such a series of tests. A circumferential scratch or tool mark on a specimen may, by acting as a local stress raiser in reversed bending, provide a starting point for a fatigue crack and thus bring about a misleading indication of the fatigue resistance of the metal of which the specimen is made. Final surface finishing should therefore comprise an operation which would ensure the absence of any circumferential scratches, however light, and so that any imperfections would be only in a longitudinal direction; moreover it should produce specimens with surfaces of identical smoothness. As perfect concentricity is necessary in order to avoid vibrational stresses, the forming of proper specimens requires precision operations.

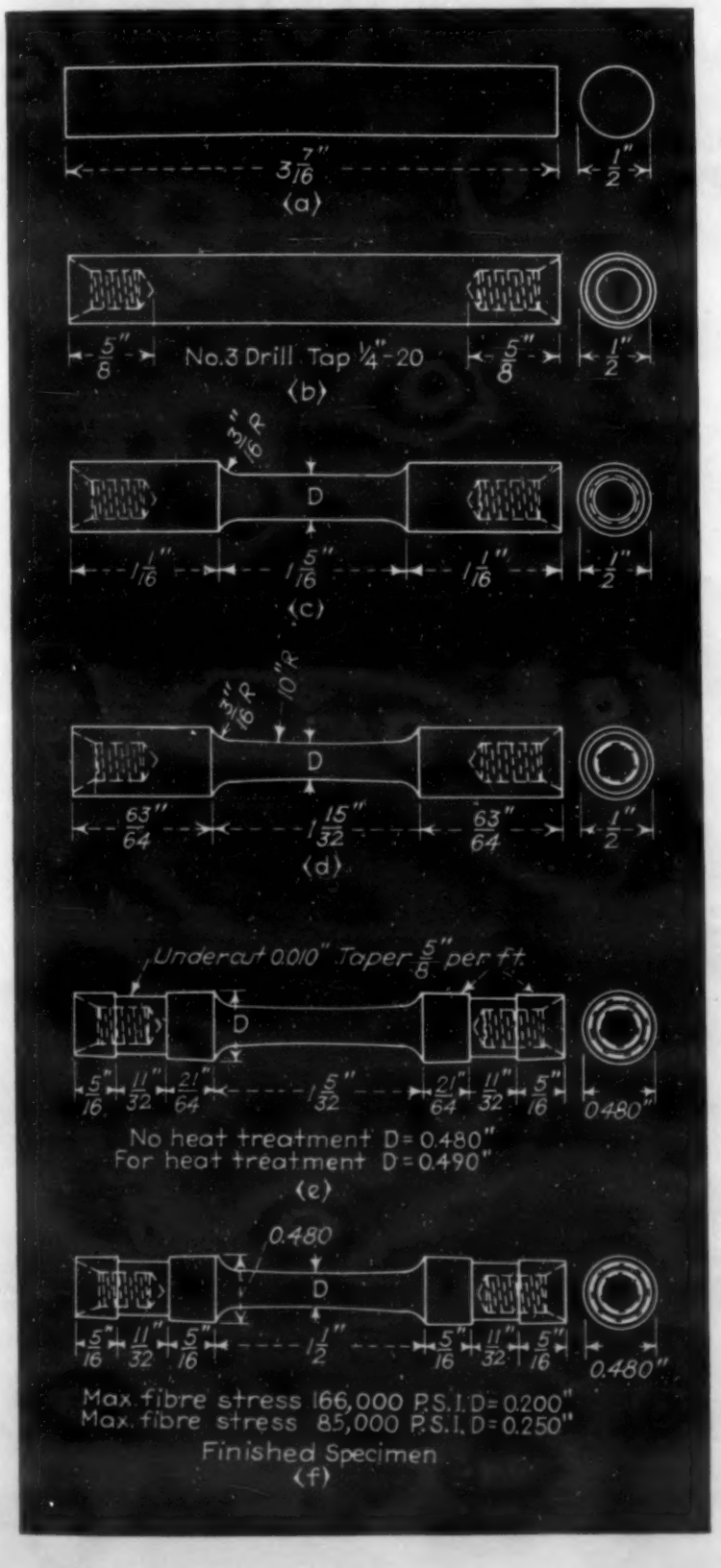
To fulfill the several requisite conditions, a procedure has been developed in this laboratory for forming and finishing specimens to be tested in the R. R. Moore type rotating beam fatigue machine. Forming operations are facilitated by the use of two special fixtures, which are readily attached to standard machines, thus avoiding investment in special machinery. A special piece of apparatus has been developed for the final longitudinal polishing of the specimen. These aids in the preparation of satisfactory specimens will now be described and illustrated

Forming Operations

The metal to be tested is cut and machined into cylinders as shown in Fig. 1(a). The ends are then center drilled to a fixed depth by means of a center drill with attached gaging collar; this allows the later plunge grinding operation to be performed without resetting for each specimen of a group. Each end is then drilled to a depth of $\frac{5}{8}$ in. by means of a No. 3 B&S gage drill [Fig. 1 (b)]; this permits easy tapping for $\frac{1}{4}$ in. USS threads, and provides ample strength for the purpose intended. Fig. 1(c) shows the center section as rough machined. The next operation is performed with a special radius turning fixture, which contains a single point cutting tool and is designed to form the center section to the contour as shown in Fig. 1(d). The use of a small cutting point prevents undue distortion of the surface layer of the specimen. The ends are then tapered and undercut, as shown in Fig. 1(e). Specimens which are to be tested without further heat treatment are finished to fit a tapered gage. Those to be heat treated are left 0.010 in. oversize on the bearing surfaces of the tapers, for removal in final grinding. A specimen which is to be tested at a hardness less than Rockwell C 40 is given the required heat treatment prior to forming; a specimen whose final hardness is to be Rockwell C 40 or above

is heat treated before the final grinding operation. The final form of the specimen, as shown in Fig. 1(f), is produced by plunge grinding the center section using a preformed grinding wheel. This is an "Aloxite" wheel, made by The Carborundum Co., 60 grit, N grade, 27 bond. A special fixture is used to dress this wheel; owing to the uniformity of contour of the specimens, very little redressing is required.

Fig. 1. Diagram to Show the Sequence of Operations in the Machining of the Specimens.



Polishing

The polishing of the center section is done in two steps. First, the specimen is polished circumferentially with No. 500 emery paper while being rotated. Next it is polished in the special longitudinal polisher, and after a suitable period it is removed, washed with gasoline, followed by carbon tetrachloride, and examined under a binocular microscope at a magnification of 60 dia. If no circumferential scratches are visible, the specimen is coated with a medium (such as a good grade of lanolin) to protect against corrosion or accidental scratching. It is current practice to leave this coating on during the test, thereby minimizing the influence of a variable laboratory atmosphere.

Special Equipment

The radius turning fixture, shown in Fig. 2, consists of a tool bit held in a pivoted head with the cutting edge of the bit extending 3/16 in. beyond the axis of the pivot. This head is carried on an arm which is pivotally mounted on the base of the fixture. The distance from the cutting edge of the bit to the axis of the pivoted arm is 10 in. Both arm and head are rotated by means of a worm and segment. The base is arranged to fit the compound rest of a 14-in. lathe, and contains means for gaging the travel of the arm and head.

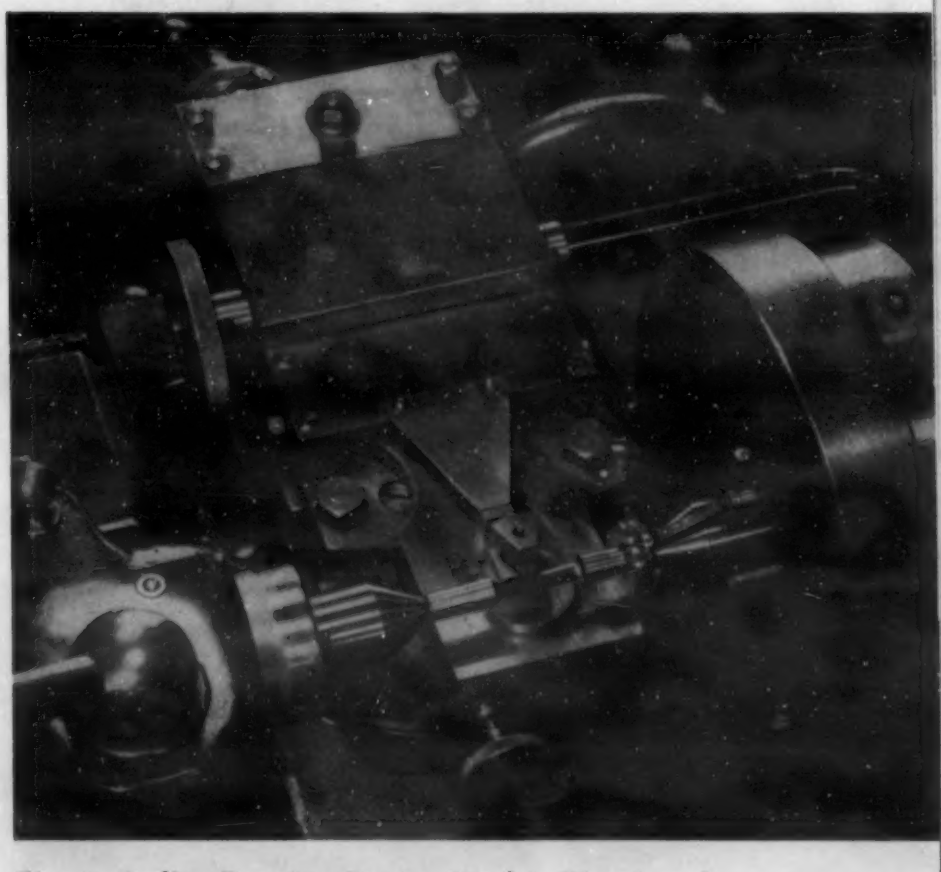


Fig. 2. Radius Forming Device Used in Turning the Central Part of the Specimen to the Desired Contour. (Fig. 1d)

Fig. 3. Truing Device Used to Dress the Grinding Wheel to the Proper Contour.

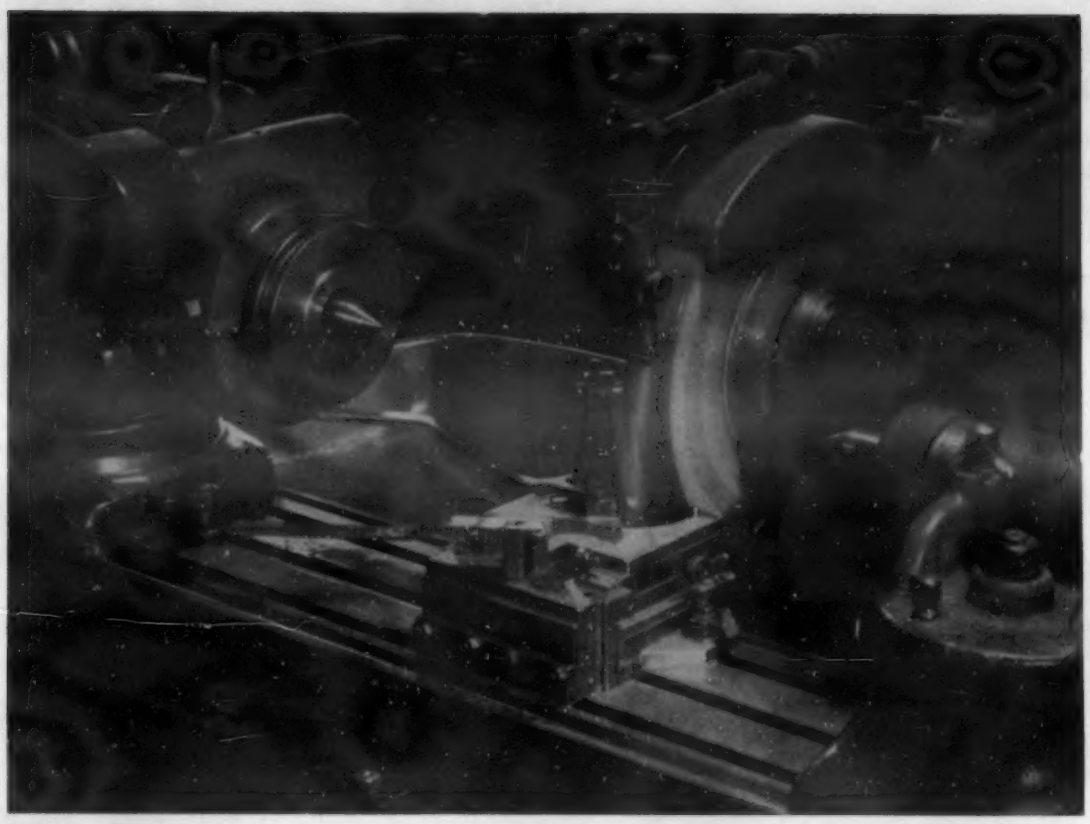
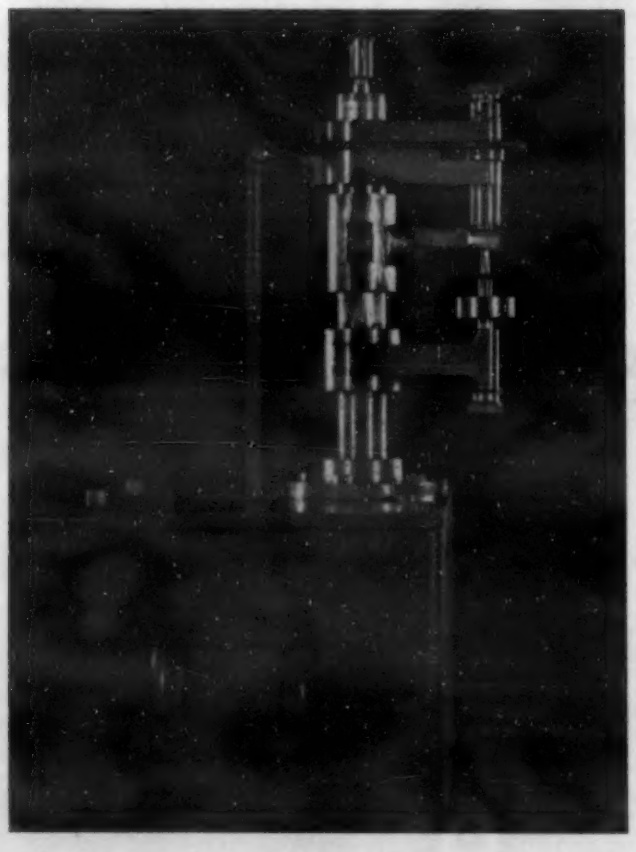


Fig. 4. (Below) Device for the Final Longitudinal Polishing of the Central Part of the Specimen.



The fixture (Fig. 3) for dressing the grinding wheel, is similar in principle to the turning device. A diamond is mounted pivotally on a carrier attached to a sub-base with the cutting edge 3/16 in. back of the axis of the carrier. The sub-base is guided by a cam to follow a 10-in. radius on the main base, which clamps to the table of the grinder. Movement of the carrier around its axis is controlled by a simple handle, and the sub-base is actuated by means of a handle and gearing, gage stops being provided for both motions.

Fig. 4 shows the longitudinal polishing apparatus which comprises a reciprocating arm and means for rotating the specimen. The base contains the motor and driving mechanism and supports a tube carrying three radial arms. The top arm, which forms the bearing for the specimen holder, is rigidly fastened to the tube; the bottom arm has a spring loaded center and is free to swing horizontally to facilitate mounting and removing the specimens. The center arm is attached to a push rod working inside the tube, and this assembly is moved up and down by means of a cam drive. This center arm carries a pair of pivoted, self-aligning fingers to which the polishing pads of felt are attached. A spring between the fingers forces them into contact with the specimen. The length of stroke is such that the entire center section of the specimen is traversed by the pads; the speed is approximately 250 strokes per minute. The specimen holder is rotated slowly, at about one revolution per minute, by means of a chain drive and reduction gearing. A thin paste of machine oil and 320 mesh emery powder is used as the polishing medium, and this mixture is applied manually to the pads. The time required to polish depends upon the composition and hardness of the specimen and proved to be the same for each of a group of similar specimens; it varies between 20 and 30 min., during which time little attention is required. Machine polishing has the decided advantage of providing automatically a duplication of surface finish, which is better than that produced by tedious hand work with its attendant risk of difference in final surface from one specimen to another.

A new polishing machine is in the course of construction which will be capable of handling four specimens at one time.

Electron Diffraction Examination of Metals

by M. L. FULLER

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Electron diffraction offers the metallurgical engineer and metallurgist a new tool for solving a variety of problems associated with metal surfaces. Because the penetrability of the electron stream is much less than that of an X-ray beam, electron diffraction may be used for examining surface formations, structures or effects too thin to be studied by X-rays. Some of the many possible practical applications of electron diffraction were discussed in an article in the March and April, 1938, issues of METALS AND ALLOYS, *particularly those relating to corrosion and oxidation problems.*

A wider knowledge of the design and construction of the equipment used in electron diffraction studies is required before this new metallurgical tool can enjoy the broad application it deserves. This article presents, to the best of our knowledge, the first detailed description of an electron diffraction camera published in a metallurgical magazine.—The Editors.

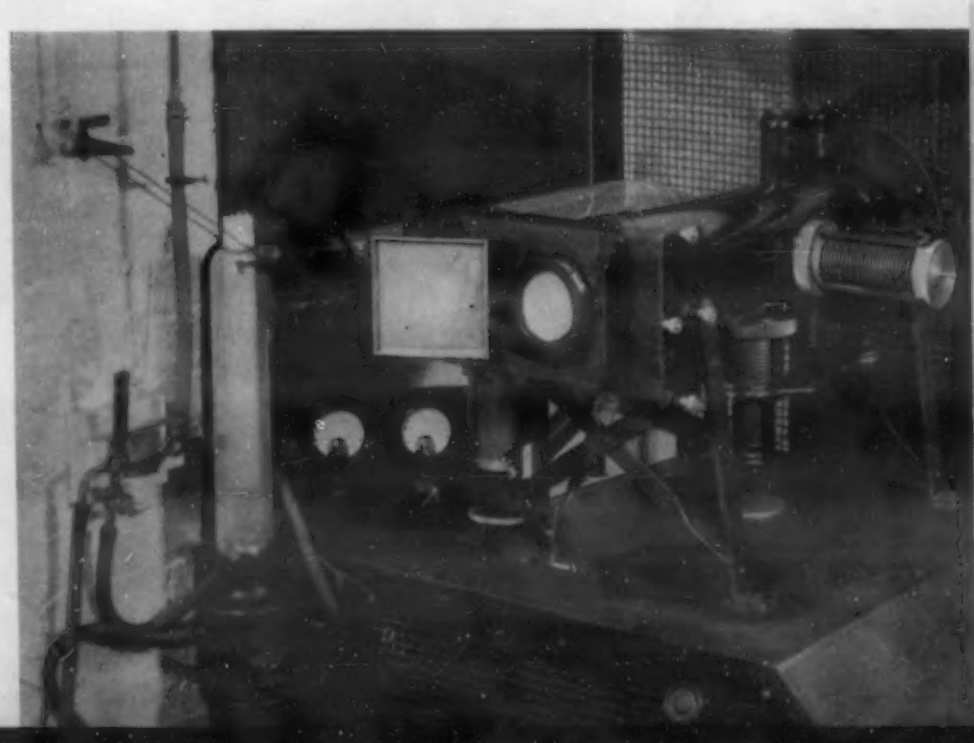
THE PURPOSE OF THIS ARTICLE is to describe an electron diffraction apparatus and to indicate briefly its application to metallurgical problems. Like X-ray diffraction, electron diffraction is a research tool for the examination of materials. Because electron diffraction has been in use only during the last decade, its possibilities in industrial research have barely been touched. It is not unreasonable to predict, however, that it will some day take its place with the microscope and X-ray diffraction apparatus as a necessary technique in the prosecution of fundamental and applied metallurgical investigations.

Many metallurgical laboratories now carry on an active program of investigation employing the X-ray diffraction method of examination. It is widely applied in the identification of alloy phases, in the determination of phase equilibrium diagrams and in the study of the fundamental processes of deformation, heat treatment, and aging of metals and alloys. X-ray and electron diffraction are similar in principle, each depending upon the interaction of radiation with the ordered space array of atoms that constitutes crystalline material. In the former case, the radiation is a beam of X-rays and, in the latter case, the radiation is a stream of moving electrons of uniform velocity behaving as a wave type radiation.

In the examination of massive specimens, both methods are limited to the examination of surface layers, with the important distinction, however, that whereas an X-ray examination may include a surface layer of the order of 0.01 mm. in depth, an electron examination is limited to a depth of the order of 1/10,000 of this amount or 0.000001 mm. Thus a surface structure, too thin to be apparent to X-ray or microscopical examination, may be studied by electron diffraction. The investigation of the outermost surface layers of a specimen is, then, the particular field of application of the electron diffraction method.

While surface examination will probably prove to be the more useful mode of study, the investigation of very thin films, thin enough to transmit the electron beam, will also be of value. Such films, of the order of 0.00005 mm. thickness, are well suited to this technique of examination, and, moreover, are too thin to yield photographable X-ray diffraction effects. Later in the paper, specific examples of both the surface "reflection" and transmission types of electron diffraction photograms will be given. Before presenting a discussion of selected applications of the electron technique, a description of the apparatus will be given. This will aid the general reader toward an understanding of that which is involved experimentally and it is hoped will be of value to

Fig. 1. Electron Diffraction Camera.



those who may contemplate the construction of a camera for their own use.

The Electron Diffraction Camera

In X-ray diffraction apparatus the source of radiation and the camera are separate units but in the electron diffraction apparatus the source of the electron radiation, the specimen to be examined and the photographic element are contained in one evacuated chamber. The camera described in this paper was designed and built in this laboratory and has been in continual operation since 1936. The writer acknowledges with deep appreciation the privilege of inspecting the apparatus of Dr. L. H. Germer at the Bell Telephone Laboratories and that of Dr. Raymond Morgan at the University of Pennsylvania before designing this unit. Features of each of these successfully operating cameras were incorporated in the present camera and mention of these features will be made in the following account. The electron source is of the hot filament type and the velocity or effective wave length of the electrons is controlled by supplying the cathode ray tube with uniform voltage direct current, usually of the order of 40 kv. Another type of electron diffraction camera employs as electron source a cold cathode with a gas discharge. The camera described by Finch and Quarrell[5] is an example of this type.

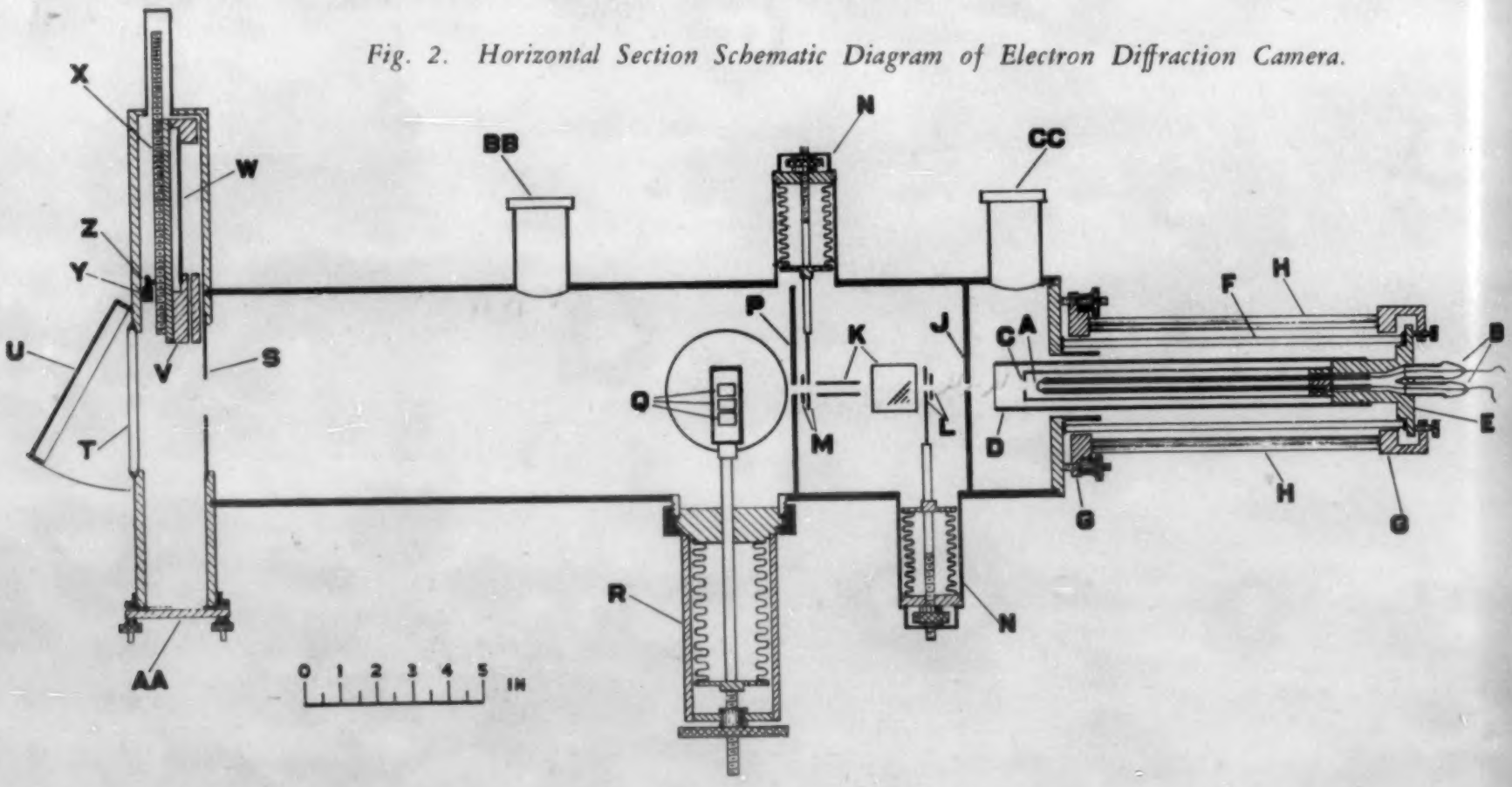
A photograph of the electron diffraction camera is shown in Fig. 1 and a schematic diagram is shown in Fig. 2 as a section in the horizontal plane.

The Filament Assembly

The designs of the filament assembly and of the collimating system follow very closely those of L. H. Germer[1] of the Bell Telephone Laboratories. The filament (A) is a simple U-shaped tungsten wire of 0.008 in. diameter. The filament is mounted on two nickel rods which are connected to the electrical circuit outside the camera through two lead-in seals (B). Surrounding the filament is a copper tube over the end of which is a nickel plate (C) having a rectangular slit 0.025 in. x 0.080 in. cut into it. This tube and slit are insulated from the filament and are maintained at 45 volts positive with respect to the filament. Surrounding this tube is a second copper tube (D) which serves as a focusing sleeve and may be adjusted to its optimum position in the initial setting up of the camera, after which no further adjustment is necessary. The sleeve is in electrical contact with the slit tube (C). The entire filament and focusing assembly is mounted on a brass head (E), the latter being connected to the source of negative high voltage direct current.

Insulation of the negative filament assembly from the positively grounded body of the electron camera is by means of a heavy walled glass cylinder (F). Rubber gaskets at the ends of the glass insulator maintain the vacuum seal effectively. The insulator filament assembly is bolted to the body of the camera by means of two brass rings (G) separated by three hard rubber rods (H), and three sets of clamps and thumb nuts. For simplicity of illustration, the diagram shows only two of the three sets of rods and

Fig. 2. Horizontal Section Schematic Diagram of Electron Diffraction Camera.



clamps in the plane of the section, whereas in the actual apparatus the three are situated at 120 deg intervals about the circumference of the insulator. The replacement of a filament, involving the disassembly and assembly of the insulator filament unit, is very easily done. The direction of the cathode ray may be adjusted by tilting the filament assembly one way or another by varying the pressure exerted by the several clamps on the rubber gaskets. This type of insulator assembly has been employed for several years in this laboratory on an X-ray tube as well as on the electron camera and it has proved to be very satisfactory.

The Collimating System

The grounded body of the electron camera is built upon a brass tube six inches in diameter. The anode plate (J) has a 3/16 in. square opening. The collimating chamber consists of two pairs of electrostatic deflector plates (K) and two pairs of slits (L) and (M). The two deflector plates of a pair are separated 1/4 in., one plate is grounded and the other may be connected through a glass lead-in seal to an external voltage for altering the course of the beam, if necessary. The size of the slit openings are greatly exaggerated on the diagram. The slits are rectangular with platinum jaws, the rectangular openings of the members of each pair having their long directions at right angles to one another. Each slit is mounted separately and may be adjusted by translation perpendicular to the axis of the body of the camera and perpendicular to the long direction of the slit. This manipulation is made by means of a Sylphon bellows (N), two of which are shown on the schematic diagram, Fig. 2. The two not shown operate vertically with respect to the plane of the diagram and may be seen on the top of the camera in the photograph of Fig. 1. Each slit opening is 3/16 in. long. The two pairs of slits, horizontal and parallel to the filament, are 0.0035 in. wide; the two pairs, vertical and perpendicular to the filament, are 0.0035 in. wide at (L) and 0.0080 in. wide at (M). The Sylphon bellows assembly for each slit is made vacuum-tight with Picein wax and may be removed for cleaning or adjustment of the slits. The exit end of the collimating chamber is partially closed with a baffle plate (P), having a 3/16 in. sq. opening.

The Specimen and Plate Holders

The complete details of the specimen holder are not shown on the diagram of Fig. 2 but will be described later. The holder is usually arranged to hold three "reflection" specimens at (Q). The bellows mechanism (R) is arranged to move the specimens horizontally and perpendicular to the axis of the camera so that the three specimens may be successively placed in the path of the beam. The latter

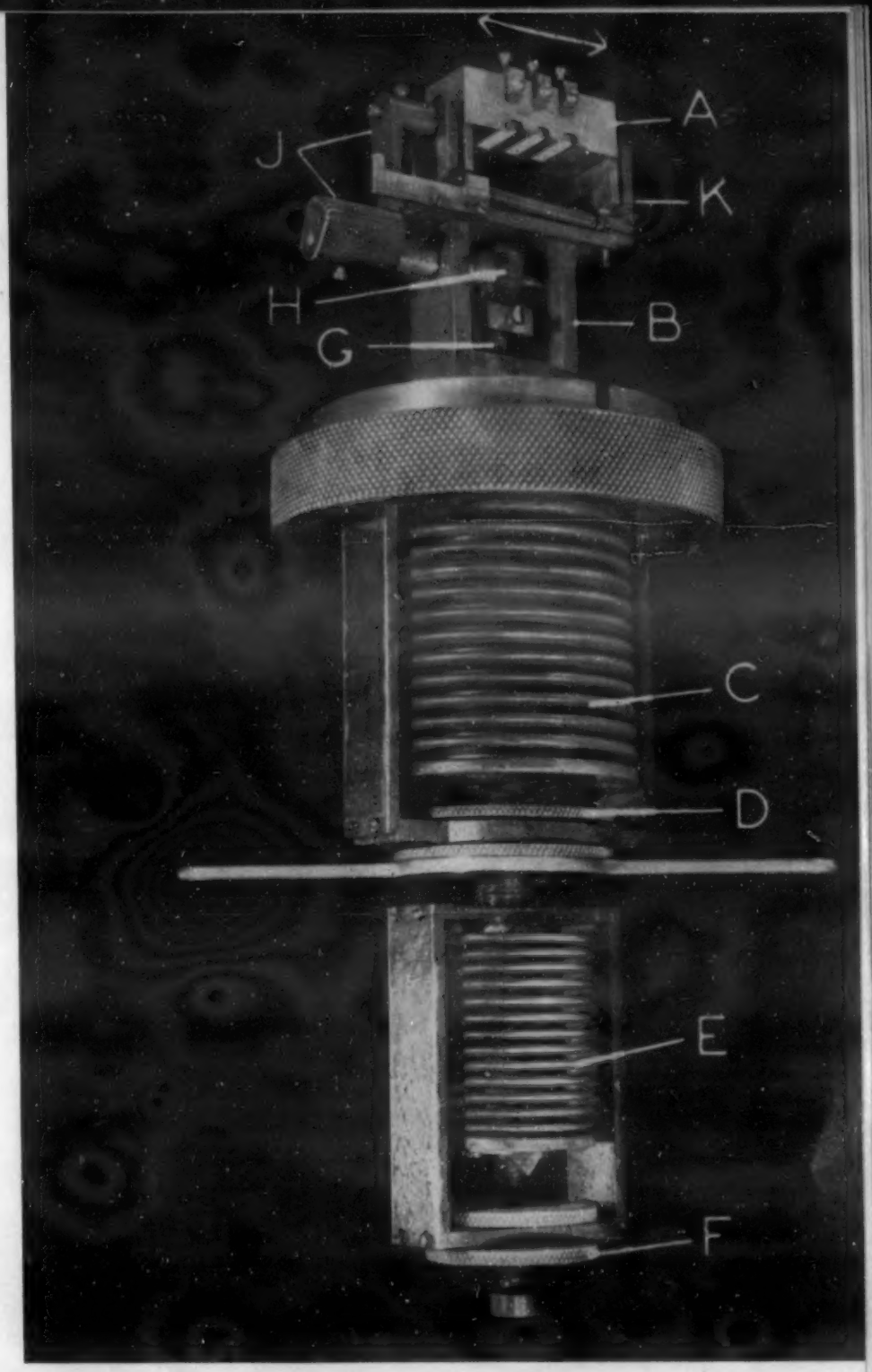
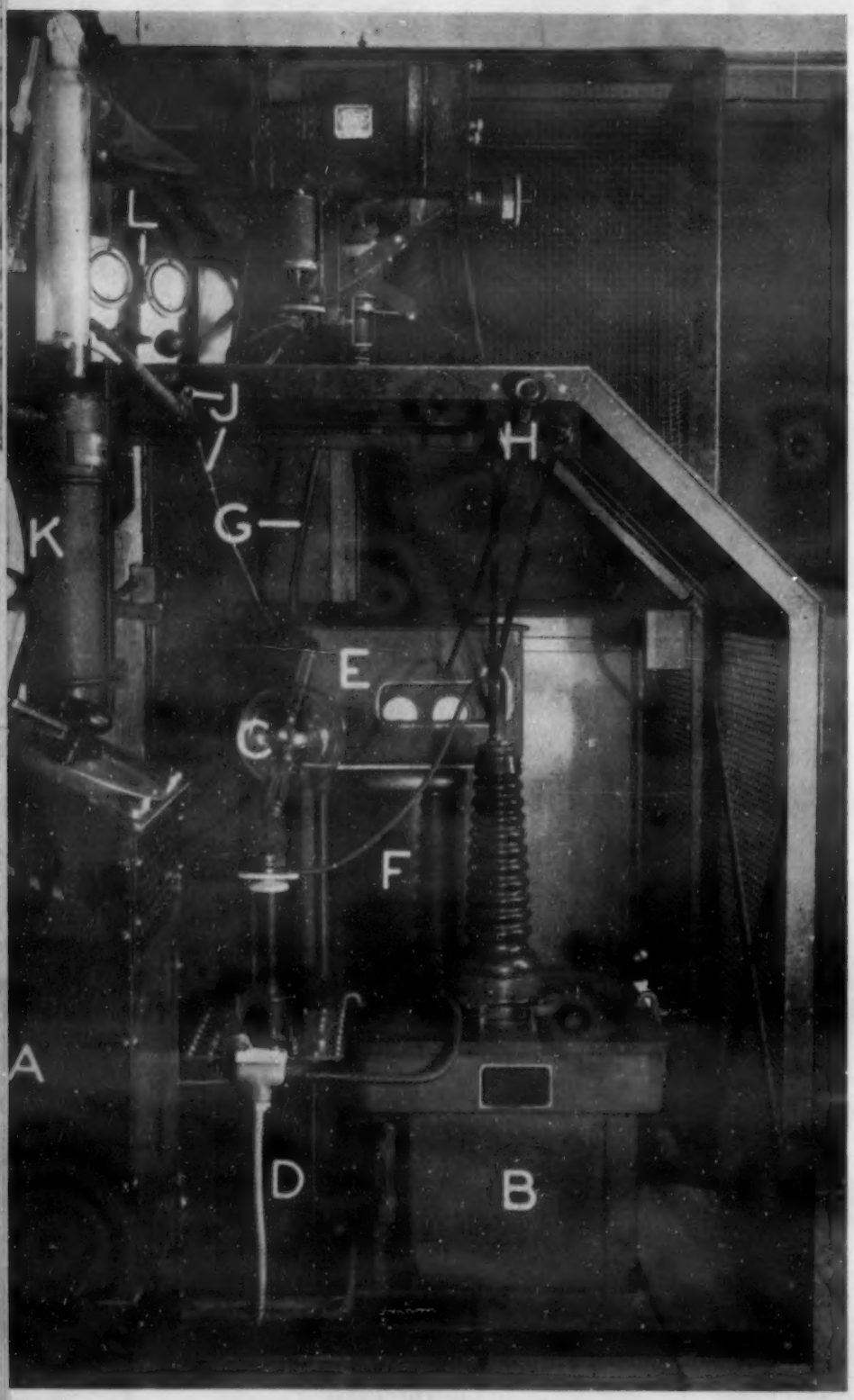


Fig. 3. Specimen Holder for Electron Diffraction Camera.

mechanism (R) is sealed by rubber gasket to the body of the camera.

The plate holder chamber was designed in a manner similar to that of Morgan and Smith[2] at the University of Pennsylvania. A rectangular diaphragm (S) separates the main body of the camera and the plate holder chamber. The diaphragm limits the area of photographic plate exposed at any one time. It is readily removable so that other diaphragms may be inserted. The fluorescent screen (T) is a 4-in. diameter "sight glass" coated with specially prepared willemite. The willemite was supplied through the courtesy of Dr. H. W. Leverenz[3] of the RCA Victor Laboratories. During the photographing of a diffraction pattern, the cover door (U) is closed to prevent fogging of the plate from external light. The plate holder proper (V) is constructed of brass in such a manner that it may be removed from and inserted into the camera in daylight as with an ordinary photographic camera. The photographic plate (W) is slow

Fig. 4. Electrical Apparatus for Electron Diffraction Camera.



and fine-grained, in size 4 in. x 5 in. Affixed to the lid of the plate holder is a long rack (X) by means of which the plate holder may be caused to traverse the entire length of the plate holder chamber. Actuation is by means of a large bellows, not shown on the diagram, but clearly shown on the photograph of Fig. 1. A second rack (Y) is attached axially to the bellows. This rack operates at right angles to the rack on the plate holder. The two racks are connected by a pinion (Z) and large gear (not shown) in 1:10 ratio. With the plate holder in the position shown on Fig. 2, the camera is adjusted for operation with the fluorescent screen. In taking a photograph, the plate holder is moved by the bellows, rack and gear mechanism until it is opposite the aperture (S). A series of exposures may be taken on each plate of the several specimens. The plate holder may be removed for dark room processing from the opening covered by the metal cover plate (AA). This cover plate is made vacuum-tight with a rubber gasket. Large diameter vacuum ports are provided at (BB) and (CC).

A photograph of the specimen holder is shown in Fig. 3. The scale of the photograph may be judged approximately by the fact that the overall length of the assembly is 11 in. Its position in the camera is indicated in Fig. 2 at (Q) and also on the photograph of the assembled camera, Fig. 1.

The Specimen Carrier

The specimen carrier (A) on Fig. 3 holds three "reflection" type specimens, appearing on the photograph as three white strips resting against slots at the bottom of the carrier. The support (B) for the specimen carrier is affixed to the bottom plate of the bellows (C). Rotation of the wheel (D) causes the bellows (C) to extend or compress, thus causing a vertical adjustment of the specimen carrier assembly, an adjustment required to bring the reflection specimens to the optimum position in the path of the electron beam. An adjustment of the angle the specimen surface presents to the oncoming electron beam may be made independently of the vertical adjustment. The bellows (E) actuated by the wheel (F) raises and lowers a rod (G). The rod (G) is attached at its lower extremity to the bottom of the bellows (E) and at its upper extremity to a rack (not visible on the photograph) that actuates the pinion (H) and thus the sector gears (J). Manipulation of the wheel (F), thus causes an oscillation of the specimen carrier (A) about an axis in the surface of the specimens. For lateral adjustment of the specimen carrier in order to bring the several specimens successively into the path of the electron beam, the specimen carrier (A) may be moved with respect to its support (B) in the direction indicated by the double arrow on Fig. 3. This is accomplished by a "pusher" operated by the bellows (R) on Fig. 2, that pushes against the end of the specimen carrier at (K). The reverse translation of the specimen carrier is accomplished by two springs, one of which is shown on Fig. 3, when the "pusher" on the bellows (R) of Fig. 2 is retracted.

Vertical adjustment, lateral translation and rotation of the specimen may thus be carried out independently, so that one adjustment does not change either of the other two.

Electrical Apparatus

The general layout of the electron diffraction unit,

particularly the electrical apparatus, is shown in Fig. 4. The electron camera is mounted on a grounded iron table at a height such that the fluorescent screen is at the eye-level of the operator. All of the high voltage equipment and connections are beneath or behind the table and the whole of the high voltage equipment is enclosed with a grounded metal screen. (One screen was removed for the purpose of taking the photograph of Fig. 4). The supply voltage, stabilized by a voltage regulator at 115 ± 1 volt a.c., energizes an auto transformer (A) by which the primary of the high voltage transformer (B) may be supplied with a selected voltage. The high voltage transformer rests in a pit two feet below the floor level. The alternating current output of transformer (B) is supplied to the cathode of a rectifier tube (C). A filament transformer (D) is provided for the rectifier tube. The anode of the rectifier is connected to a copper box (E), the latter being mounted upon a capacitor (F) of 0.01 microfarad capacity and connected to the ground. A connector (G) from the copper box to the cathode of the electron camera supplies the camera with substantially uniform voltage, direct current. Within the copper box (E) are a 4-volt storage battery for heating the filament of the electron camera, a 45-volt dry battery for energizing the slit-plate mounted over the filament (C) in Fig. 2, ammeters for measuring filament current and filament emission, coarse and fine adjustment rheostats for controlling the filament current, and switches for the filament heating circuit and the filament emission circuit. The copper box (E) and its contents are, of course, at the high voltage so that the rheostats and switches must be manipulated by means of insulated rods at (H) and a string at (J).

The vacuum is maintained with a four-stage steel mercury diffusion pump (K, in Fig. 4), backed by a rotary oil pump, not illustrated. The mercury pump is surmounted by a glass freezing trap, employing dry ice and acetone, thence two 45-millimeter diameter glass tubes lead to the vacuum ports of the electron camera. The camera can be evacuated to a working vacuum in ½ hr. or less. The panel (L) is the indicating and control unit for a thermocouple vacuum gage.

The entire unit is very compact, occupying a floor space of 3½ by 5½ ft., excluding the auto transformer. The high voltage equipment is adequately guarded so that it is shockproof. The equipment is placed in a room that can be darkened when viewing the fluorescent screen.

(To be concluded)

References

[1] L. H. Germer. *Review of Scientific Instruments*, Vol. 6, 1935, page 138.

[2] R. Morgan & N. Smith. *Review of Scientific Instruments*, Vol. 6, 1935, page 316.

[3] H. W. Leverenz. *Journal of the Optical Society of America*, Vol. 27, 1937, page 25.

Editorial

(Continued from page 67)

plant, and the extent to which such control has contributed in each case to unusual quality of finished product or to economical operation.

We believe that the majority of those general industrial factors that fabricate, treat and use metals could materially simplify their metal-handling problems, from plant layout through the purchase of metal working and heating equipment down to the assembly of the finished product, be it airplane or machine screw, if they would consider the separate operations each performs on metals not as disconnected jobs without apparent relation to each other, but as a potentially co-ordinated series of unit metallurgical processes. The processes are actually tied together by the metal passing through each of them even though this may not be evident because of frequent barriers and diversions in the form of "product control" systems, stock rooms, tool rooms, order clerks, inspectors, millwrights, and power failures. Why not complete the tie-up with unified metallurgical control of these metallurgical engineering operations?

The unit operations in metallurgical engineering have been with us for a long time, but by and large, (with the exception of the monthly reminder given by this magazine in the classification of its CURRENT METALLURGICAL ABSTRACTS) they have wandered through industry without recognition as such. Rolling, drawing, machining, annealing, case hardening, heat treating, pickling, electroplating, welding, cutting and many others are all of the same breed—they are unit metallurgical operations—and industry can but profit by regarding, organizing and controlling them as a completely co-ordinated whole. Even in specifying equipment for performing primarily mechanical operations, such as machine tool work, the mechanical engineer can be more certain of a wise selection if he consults the metallurgical engineer, who may know little about the construction and operation of the machine, but who knows a good deal about the machining properties of the steel, the most suitable cutting tool materials, and the best feeds, speeds and coolants for the combination.

Here is one of the biggest jobs the industries that use metals predominantly can do for themselves: Recognize the unit metallurgical operations for what they really are; place their control in the hands of men trained by education or experience for just such work—the *metallurgical engineers;* then watch how production problems are simplified and product quality improved.—F.P.P.

Sand Mold Ready for Closing Before Pouring. (Courtesy: Warner & Swasey, Cleveland)

Sand Control for Better Castings

Few foundrymen today can boast a business life of unbadgered bliss. The competition of other production and fabrication methods, the generally low profit margin, and the customer's insistent demand for castings of ever higher quality have furrowed the supervisory brow and grayed the managerial hair. To master his difficulties, the foundryman has done many things—"educated" designers, mechanized his handling and cleaning, improved his melting practice, investigated alloys, correlated casting conditions with service characteristics. And now he is fast learning how vital to the achievement of best results are the use of the correct type of molding sand for each job, and the application of a carefully-planned system of sand testing and control.

This article, based on a paper presented before a regional meeting of the Amer. Foundrymen's Assoc., discusses in practical fashion just how the type of sand affects casting quality for various metals, and what simple methods are available today for testing, controlling and standardizing the grades and quality of foundry sand used.—The Editors.

by H. W. DIETERT
President, Harry W. Dietert Co., Detroit.

THE IDEA THAT THE NATURE and properties of sand mixtures powerfully affect the quality of finished castings is older than the oldest foundryman. But it is only recently that we have learned what the amount and incidence of these effects are, and only very recently that we have applied systematic sand testing methods to the economically important problem of controlling molding and core sands, and, with them, the quality of our castings.

Sand testing as ordinarily performed is a simple, clear cut process, with two main objectives: To *measure* certain physical properties of the sand, and then to *hold* the sand to a predetermined set of physical property values. The modern foundry finds it profitable to control four sand variables—moisture, permeability, strength and grain fineness; these generally exert the greatest influence on the finished casting. Also important to the foundryman are mold hardness and flexibility, properties which he has learned to develop through judicious handling of his sand. All these properties are somewhat interrelated, as we shall see, but basically efficient sand control depends on a knowledge and standardization of three characteristics of any given sand mixture: Moisture, grain size and clay content.

Moisture

The most important physical property to control in a molding sand is the moisture. Although moisture tests may be made by the practical man in the foundry or by the technician in the laboratory, the greatest benefit is naturally secured when the moisture is checked in the foundry at the time the sand is mixed. Where this is done the man at the sand mixer checks the moisture of every batch, or say, every third batch of sand before the sand is dumped. The moisture test figure indicates to the operator the amount of water that should be added, and avoids that bugaboo of all careful processing—guesswork.

The test for moisture is very simple when a moisture teller, as illustrated in Fig. 1, is used. A weighed sand sample is placed in the drying pan and put under the funnel of the moisture teller. The blast of heated air flashes the water to steam and dries the sand, the time required in many cases being less than one minute. The dried sand is reweighed, and the loss in weight read on the rider scale of the balance as moisture percentage.

Moisture variation in a steel sand should not exceed 2 points, plus or minus, of the desired figure. For gray iron molding sand a variation of 4 points, plus or minus, of the chosen figure is permissible. In molding sand used for non-ferrous castings the moisture should be held to within five points of the best workable moisture range. The best workable moisture range of a molding sand is that at which the cleanest lifts from pattern and best casting results are obtained.

When the sand is worked with insufficient or excess moisture, the casting surface is inferior in finish to that obtained when the sand is tempered to the best workable moisture content. Molding sands for light gray iron, where permeability values of 12 to 30 are used, should be tempered to a heavy tempered condition, whereas, a sand for heavy gray iron should be worked with moisture to produce a well-tempered condition. A steel sand with moisture to give a fairly heavy tempered condition is preferable, so that the sand will set firmly when dried by the molten metal or by the mold oven.

A wet sand defect is shown in Fig. 2. The "blow" illustrated possesses the sharp jagged or irregular edge characteristic of wet sand condition.

Grain Control

The grain structure of a sand should be controlled for permeability, grain size and distribution, so as to secure a porous and flexible mold mass. Porosity gives permeability, *i.e.* the mold will vent gases and contain voids into which the sand grains can move as the mold mass rearranges itself when the casting shrinks on cooling or when the sand grains grow from the heat of the molten metal. To accommodate this rearrangement of sand, a mold should possess also a certain degree of flexibility.

The permeability of a molding sand is measured by first ramming a standard sand specimen with a sand rammer, Fig. 3, and then measuring the air flow through the rammed sand specimen with the permeability meter, Fig. 4.

For aluminum castings, a molding sand with a permeability from 10 to 15 works well. For brass, a permeability of 12 to 25 is satisfactory. Where special alloys are added to non-ferrous metal, a permeability is high as 90 is often found necessary. Permeability of sand used in the gray iron foundry can be as low as 12 for stove plate; a range of 15 to 30, with 22 the most satisfactory, is used for light castings in snap flasks made on squeezers. In gray iron jobbing, a permeability of 40 to 50 gives good results for medium floor work. A permeability of 80 to 120 will work well for heavy green sand molds, while in dry sand molding a permeability from 75 to 150 is in general use. And here is contrast: The permeability range from 150 to 200 is in general use for green and dry steel sand, but permeability as low as 60 is used for stainless steel castings.

The casting defects most commonly caused by insufficient permeability are "metal agitation" and

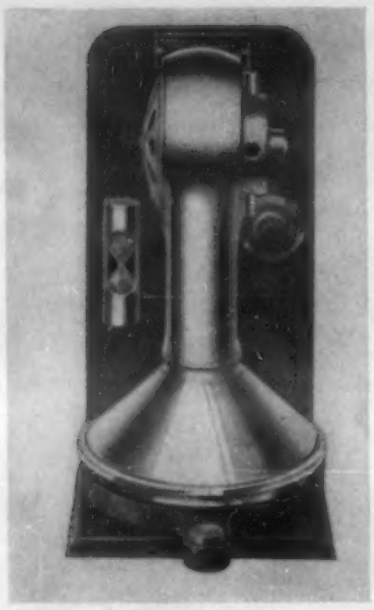


Fig. 1. (Left) Moisture Teller for Determining the Moisture Content of Molding Sand. Fig. 2. (Center) A Typical Wet Sand Defect. Fig. 3. (Right) Sand Rammer Used in Permeability Testing.

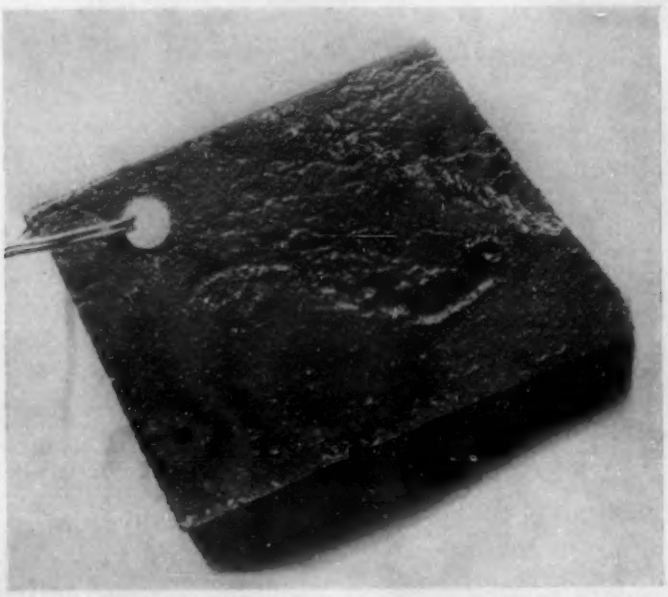


Fig. 4. (Extreme left) Permeability Meter. Fig. 5. (Left) A Characteristic Metal Agitation Defect.

Fig. 6. (Left) Surface Blow Defect Caused by Insufficient Permeability. Fig. 7. (Right) Penetration Defect, a Result of Excessive Permeability.

"blows". Where the permeability of the sand is slightly on the low side and the ferrostatic pressure in the mold is low, the molten metal may become agitated and erode the mold face. A defect of this type will appear as shown in Fig. 5. When the permeability of the sand is decidedly low, a surface blow defect as shown in Fig. 6 is obtained. Conversely, the permeability of a sand may be too high, causing metal penetration into the mold surface and giving a defect like that shown in Fig. 7. (Such defects may also be caused by low flowability.)

In certain cases a high permeability in combination with a low clay content and a certain incorrect metal condition may aggravate pin holes, Fig. 8. Such pin holes have a depressed casting surface around the pin hole crater and contain no squeezed-out metal shot. The remedy is an increase of clay content, reduction of moisture or metal correction.

The fineness of the grain structure of the sand is determined by riddling it through a stack of 11 sieves, ranging from 6 to 270 mesh, as shown in Fig. 9. The grain size distribution may be studied by noting the amount of sand retained on each sieve. Also the average fineness of the grain may be calculated by the A. F. A. grain fineness formula. Retention of sand on all the sieves in large quantity indicates poor grain size distribution, whereas retention of the sand grains on a few adjacent sieves indicates good distribution and maximum permeability efficiency.

In commercial practice the best controlled sands have a distribution such that 75 per cent of the sand grains are retained on three adjacent sieves. A sand with fairly good distribution would have 65 per cent on three adjacent sieves. Poorer distribution means fewer pores, with consequent undesirably low mold flexibility.

Clay Control

The clay content of a molding sand should be controlled because of its effect on the green strength, dry strength, flowability and sintering characteristics.

The green strength of a sand is determined preferably by measuring the load that an A. F. A. standard sand specimen will support when tested in a machine such as that illustrated in Fig. 10. The dry strength of a sand is measured in the same manner except that the sand specimen is baked and allowed to cool before testing for strength. Dry strength tests may also be made at various selected temperatures up to 3000 deg. F.

The green strength of a molding sand is largely controlled by the quantity and quality of clay bond present in the sand; the clay bond also affects the refractoriness or sintering characteristics of the sand, a high clay content lowering the refractoriness. It is essential to know the quantity of clay bond present. This is determined by first scrubbing the clay free

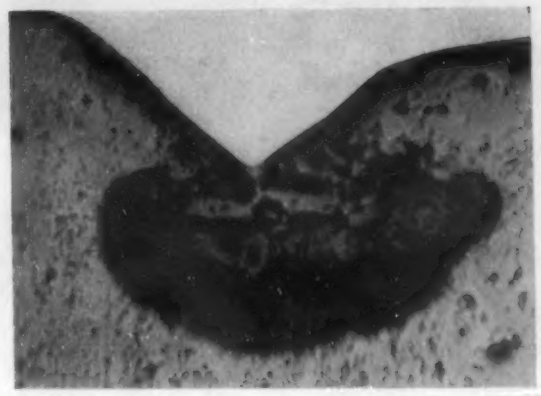


Fig. 8. Cross Section of a Pin Hole Defect Originating from a Combination of Sand Conditions.

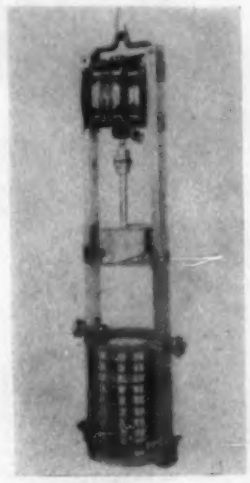


Fig. 9. Sieves in Sifter, Used for Determining Grain Fineness and Distribution.

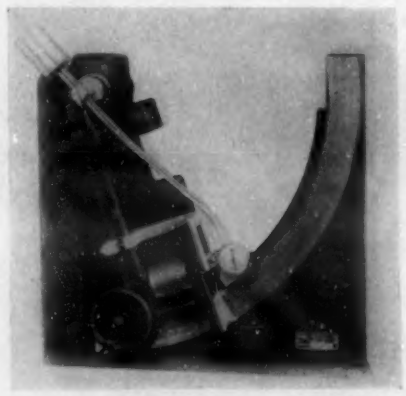


Fig. 10. Universal Sand Strength Machine.

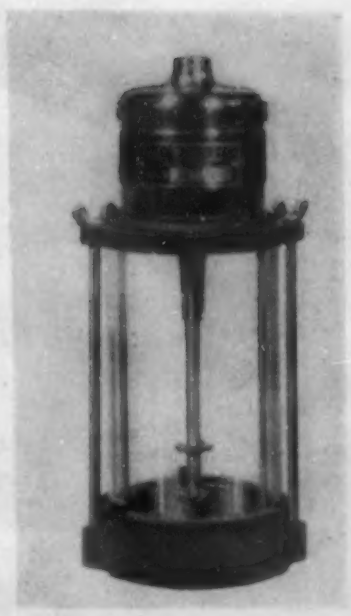
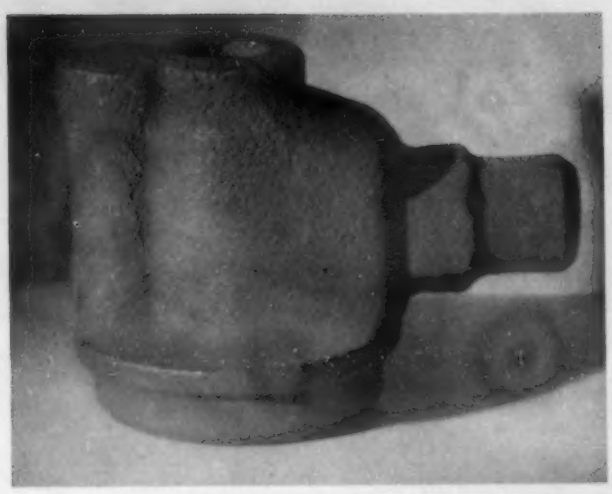


Fig. 11. Rapid Sand Washer, Employed in Checking Clay Bond.



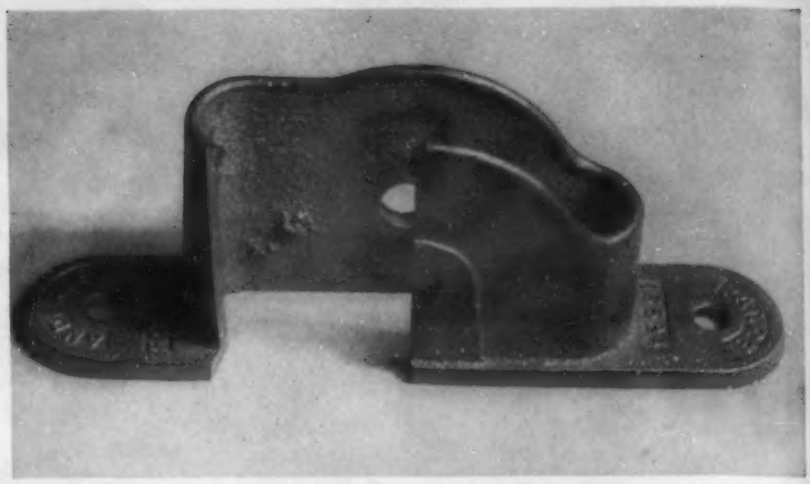
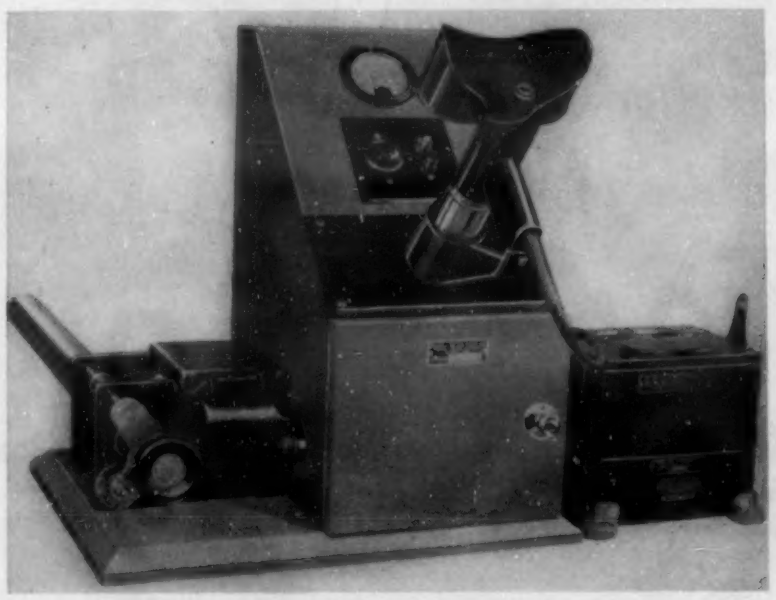


Fig. 12. (Left) Typical Wash Defect Caused by Low Green and Dry Strength. Fig. 13. (Center) A Blow on Casting Surface Resulting from Excessive Mold Hardness. Fig. 14. (Right) A Penetration Defect Caused by Insufficient Ramming in a Restricted Part of the Casting, and possibly Aggravated by Low Flowability.



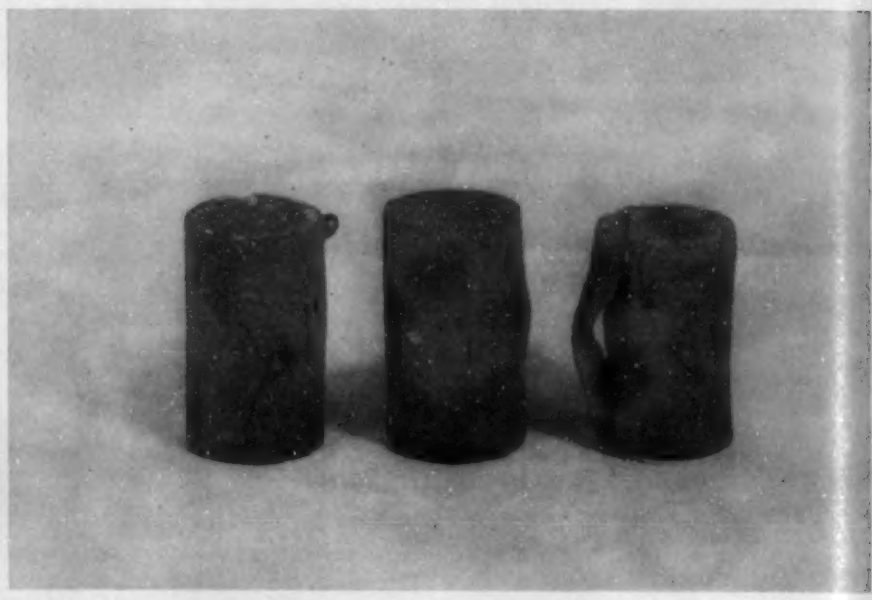


Fig. 15. (Left) Sintering Tester, Used to Measure the Temperature at which Sand "Burns On." Fig. 16. (Right) Sand Samples, after Being Subjected to a Temperature of 2500 Deg. F.

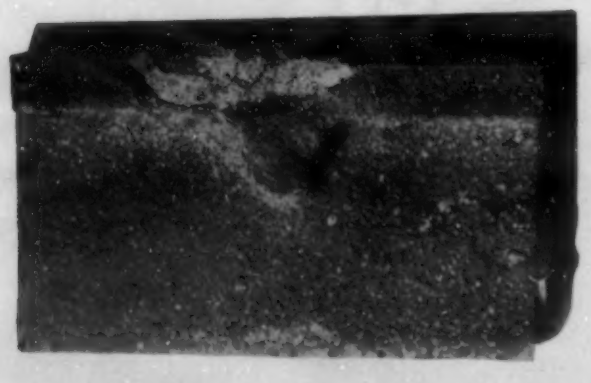


Fig. 17. A "Scab" Defect on Surface of Casting.

Fig. 18. A "Rat Tail" Defect on Casting Surface.

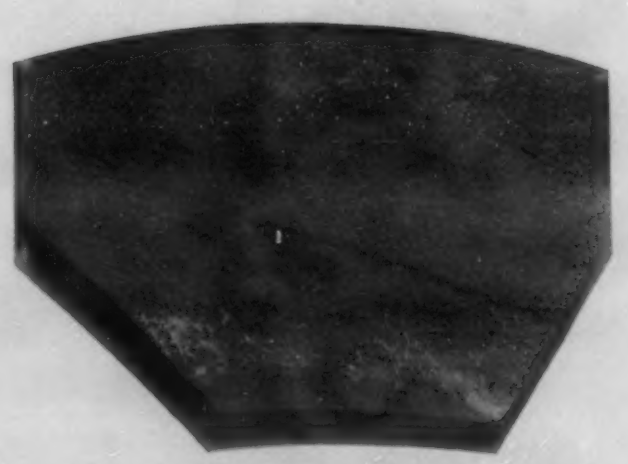
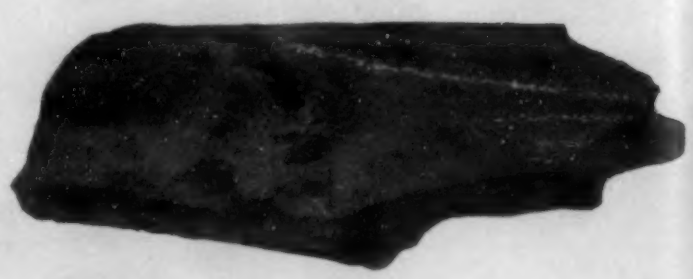


Fig. 19. A "Rat Tail" and "Scab" Defect Appearing at the Same Place.



from the sand grains with the rapid sand washer, Fig. 11. After the scrubbing operation the clay is siphoned from the sand grains in the wash bottle of the washer; an insufficient green strength will cause the sand to drop out of the mold easily.

Insufficient green strength is also usually associated with low dry strength. A low dry strength will allow the metal to erode the mold surface because the sand does not set firmly when dried by the molten metal. Such defects may have an appearance like that shown in Fig. 12. Excess green strength, which usually is accompanied with high dry strength, will set the sand in the mold so firmly that hot cracks occur on the castings made from quick setting metals. The mold will also shake out hard when molds are left standing as they usually are in a floor foundry. As one may surmise, a high dry strength also causes the mold surface to set to a brittle mass without sufficient flexibility to accommodate the volume change of the sand on heating.

Mold Hardness

A mold is rammed to secure a mold hardness that will allow the clay bond in the sand to impart to the molding sand the desired green strength, dry strength, permeability and porous structure. The mold must have sufficient hardness to support the molten metal load and yet be flexible to a limited extent.

The casting defect—specifically, a "blow"—shown in Fig. 13 was caused by ramming a sand with slightly too much moisture to a high mold hardness; in this case, 75. A casting defect like that of Fig. 14, where the mold was not rammed sufficiently in a restricted part, is a "penetration" defect and may be aggravated by low flowability.

Practices that affect mold hardness greatly influence every physical property of a sand; when the hardness of a mold is changed, the sand condition in general is modified, in most cases directly or inversely in proportion. For example, when the mold hardness is doubled permeability is reduced about one-half.

Sintering

The refractoriness of a molding sand is measured with the sintering machine, Fig. 15. The temperature at which a platinum ribbon fuses to an A. F. A. standard sand specimen is expressed in degrees and designated as the sintering point. The sintering point of a sand may be increased by increasing the grain size, moisture and amount of facing material, such as sea coal, by reducing clay content, or by the selection of more refractory clay bond and cleaner sand grains.

Mold Flexibility

The subject of mold flexibility is new and of great practical importance. A mold surface should be flexible to a limited extent to allow for volume change of molding sand when heated, and for metal shrinkage.

The sand grains of a molding sand may expand as much as 0.040 in. per in. of mold length when no clay bond is present. The clay bond in a molding sand contracts upon heating, particularly when temperatures are reached where the clay begins to oxidize and sinter—as low as 1600 deg. F. for some clay bond contained in sand.

In Fig. 16 are shown three sand samples. The one on the left expanded sufficiently to cause mold surface cracks. Note the iron shots that flowed out of the sand. The specimen in the center shows a buckle where the surface of the mold peeled off, because of the compressive force of the expansion of the sand grains. Not enough heat was present to cause the clay bond to contract; the resulting condition corresponds to copes of mold where "pull-down or buckle" occurs because of the heat radiated from the molten metal. When sufficient heat is present, as in corners or drags of large molds, the sand grains first expand and loosen or peel off the surface of the sand and then, as the latter becomes hotter, the clay shrinks and opens up the "peeled" section so the molten metal can flow behind a layer of sand, causing a "scab", as shown in Fig. 17. As one increases the flexibility of the mold surface by lower mold hardness, less clay content and better distribution of sand grains (less fines), the tendency to scab is lessened.

To eliminate the "rat tail" defect, Fig. 18, which is an expansion defect on a lighter casting, the sand is worked with a high enough clay content to reduce somewhat the expansion of the sand grains. The mold should be rammed as soft as possible, using a well-tempered sand.

As the heat is increased, a rat tail and scab may both show as in Fig. 19. A scab defect first goes through a "rat tail", or buckle stage, which is an expansion defect, and then through a contraction stage. The expansion of a sand is reduced by increasing the grain size, facing material, clay or cereal content or by decreasing the mold hardness and amount of fines. The contraction of a sand is reduced by increasing the facing materials or through the use of a more refractory bond.

General Usefulness

Sand control is clearly applicable, with distinct advantage, to all types of foundries, whether large or small, ferrous or non-ferrous. Sand control permits preparation of sand with desired moisture, permeability, green and dry strength, grain fineness and mold hardness values—sand that molds easier and produces better castings. And broadening of a sand testing system results in uniformity—the molders work a sand that is the same each day, without the fluctuations in physical properties that result invariably in spoiled castings.

PROPERTIES AND METALLURGICAL USES OF

Calcium

—A CORRELATED ABSTRACT by C. L. MANTELL and CHARLES HARDY • PART 2

Consulting Engineer, New York, and President, Charles Hardy, Inc., New York, Respectively

Calcium-Lead Alloys

The lead-calcium system, studied in considerable detail by a number of workers,[31] is characterized by the presence of the compounds Pb_3Ca, PbCa, and $PbCa_2$. Only alloys in the range between pure lead and Pb_3Ca are at present of commercial interest.

Calcium is an extremely active element that may produce a large effect when but a very small amount of the material is present. It has been stated[32] that lead with several per cent of calcium or barium or both may be too hard to be cut with a hacksaw. The age hardening properties of the calcium-lead alloys were investigated by Cowan and his associates of the United Lead Company in 1916.

During the War the demand for shrapnel bullets caused a shortage of antimony and antimonial lead, and the lead-barium-calcium alloys known as Frary[33] or Ulco Metal were developed. These hard alloys, containing up to 2 per cent barium and 1 per cent calcium, also found considerable favor for bearings. Investigators have studied the physical properties of lead alloys up to 2 per cent calcium for use as bearing materials. In these, large particles of Pb_3Ca cause hardening and furnish bearing surfaces. "Bahnmetall," widely used as a lining for bearings on German railways, contains about 0.75 per cent calcium with 0.5 per cent sodium and 0.05 per cent lithium.

The existence of a definite range of age-hardening lead-calcium alloys has been determined.[34] Some of these hardenable alloys possess properties which strongly recommend them for commercial uses. Compared with the 1 per cent antimony-lead alloy, which is generally recognized as one of the best cable sheathing materials, the lead-calcium alloys develop greater fatigue resistance, tensile strength, and hardness. These properties reach nearly constant values shortly after extrusion because of the rapid rate of precipitation of calcium from solid solution, and the slow rate of diffusion and agglomeration of calcium, or the compound Pb_3Ca. The 1 per cent antimony-lead alloy continues to change even after a period of years. Experiments on corrosion resistance show the lead-calcium alloys to be so little different from the 1 per cent antimony-lead alloy as to have no appreciable effect on the life of a cable.

The 0.04 per cent calcium-lead alloys show much higher endurance curves and greater fatigue resistance than the 1 per cent antimonial lead.[35]

Independent investigation[36] shows that the lead-calcium alloy containing 0.03 to 0.04 per cent calcium and cooled from an extrusion temperature of 225-250 deg. C. has a tensile strength of about 4,000 lbs. per sq. in., a dynamic strength of 1,200-1,500 lbs. per sq. in. as determined on the Moore machine at 1,700 r.p.m., and a static "fatigue" strength markedly superior to the 1 per cent antimony alloy. These properties make this alloy a superior cable sheathing. A calcium-lead alloy containing 0.02 to 0.1 per cent calcium was suggested as sheathing material for electrical cables.[37-43]

Large scale field tests of lead-calcium alloys on commercial telephone and cable lines have been conducted since 1930. They will serve as final criteria of the commercial usefulness of the alloys. The manufacture of cable sheaths for electrical lines requires so much lead that it is ranked as one of the three major consuming industries. It is interesting to note that if all cable sheath in the United States—an industry which consumes from one-fifth to one-quarter million tons of lead a year—were made of the 0.04 per cent calcium alloy, one hundred tons of calcium would be needed annually for this use alone.

Very small amounts of calcium appreciably increase the deformation pressure of high purity lead.[44] Lead calcium alloys containing 0.04 to 0.10 per cent calcium combine physical, metallurgical and electrochemical properties which recommend them as a superior material for storage-cell grids and plates.[45]

Batteries containing elements made of these alloys, when properly heat treated, should have a high efficiency and long service life. These alloys can be satisfactorily cast, rolled, welded, machined and burned.

For use in storage-cells, lead should be alloyed only with metals electronegative to it. A comparative study was made[46] of the electrochemical behavior of pure lead, a 9 per cent lead-antimony alloy (standard for many storage batteries today), and a 0.1 per cent lead-calcium alloy. On static discharge the lead-antimony alloy sulphates in approximately one-ninth of the time required for pure lead and one-tenth of the time required for the lead-calcium alloy. Anodically and cathodically, the behavior of the lead-calcium alloy is similar to that of pure lead, but quite distinct from that of the lead-antimony alloy. The indicated electrochemical superiority of lead-calcium alloy over the lead-antimony alloy was confirmed by actual battery tests. Lead-calcium cells retain their charge for far longer periods than lead-antimony cells and are more efficient. No undue corrosion of lead-calcium grids was observed. The lead-calcium alloy is recommended for storage battery grids in place of the lead-antimony alloy and for Planté type plates in place of lead.

In an investigation of the creep and fracture of calcium-lead alloys,[47] the McVetty method[48] for estimating total creep in a given time under given stress and temperature was employed in the latter phases. This work has important bearing on the design and use of electric cable sheathing. Low stresses were maintained for periods of 2 to 3 years. No absolute limiting tensile stress was found below which creep did not take place. With 180 lb. per sq. in. at room temperature, the rate was from 0.3 per cent per year for calcium-lead alloy to 0.6 per cent for commercial lead. At 150 deg. F. the variation was from 0.6 per cent for calcium-lead to 1.4 per cent for commercial lead. At room temperature and stresses of 500 pounds, yearly creeps were 5 per cent for lead, 3.2 per cent for tin-lead, 1.4 per cent for antimony-lead, and 0.8 per cent for calcium-lead, the superiority of alloys being manifest at stresses higher than those commonly met.

Deoxidizing Copper and Its Alloys

As far back as 1907 claim was made[49] for the use of calcium for deoxidation and degasification of copper. Masing[50] has described the use of calcium in producing deoxidized copper castings of high conductivity. This work was carried forward in the research laboratories of the General Electric Company and the Bell Telephone system and published in a series of papers.[51-53] Schumacher and his co-workers concluded that small additions of calcium in copper effectively deoxidize the metal without materially imparing the mechanical properties and electrical conductivity, provided the residual calcium is kept to a small percentage. The resulting deoxidized copper is not embrittled by annealing in reducing atmospheres.

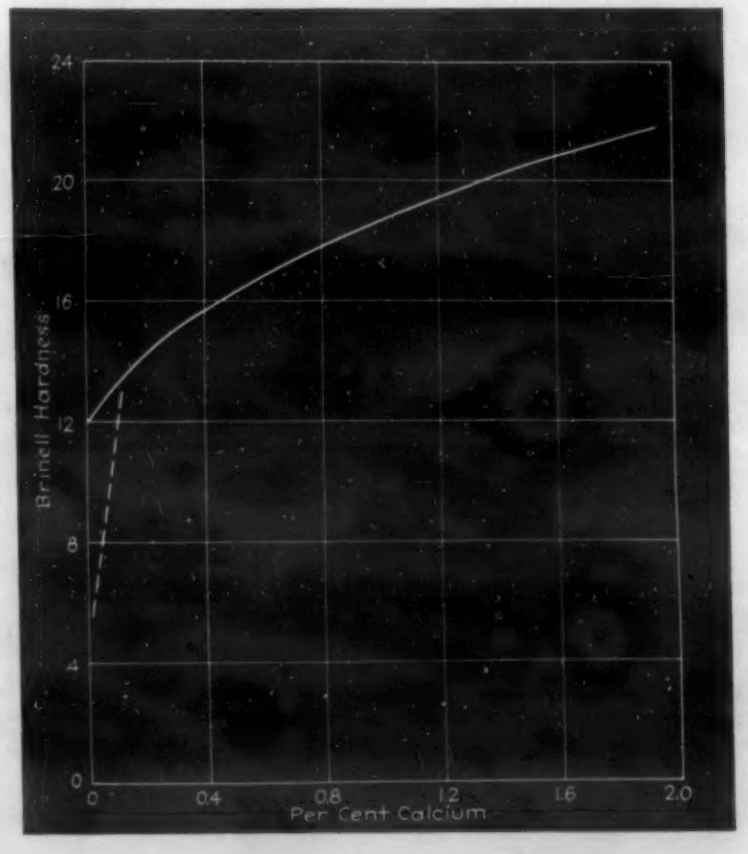
The solubility of calcium in copper is slight.[54] A study of the conductivity of copper castings[55] showed that for such castings the best deoxidizers, in reference to their effect on conductivity, were calcium, calcium boride, or boron carbide. The calcium was introduced in the form of a copper-calcium alloy.

Calcium has been shown to be preferable as a deoxidizer to other metals of the alkaline and alkaline earth groups. Small amounts of calcium give superior qualities to copper as regards resistance to embrittlement. High-calcium copper alloys are satisfactory deoxidizers. Calcium is preferred to silicon as a deoxidizer for copper, in that it affects the conductivity very little, while silicon affects it markedly.

High-calcium copper alloys, now commercially available, can be produced in brittle forms which are readily powdered or ground. Their manufacture involves the melting of copper and the addition of calcium to the melted copper. Calcium has a lower melting point than copper, and when added to the molten metal, readily melts and alloys.

Calcium is an excellent deoxidizer for high strength tin bronzes. It is also an effective reagent for suppressing lead segregation in tin-free high-lead bearing bronzes with lead contents of 30-50 per cent, with copper 70-50 per cent.[56] In copper-lead mixtures, lead segregation takes place very rapidly and is excessive in alloys containing 30 per cent or more lead.

Effect of Calcium Content on the Hardness of Lead-Calcium Alloys.



These mixtures are very useful for bearings in engines operating at high speeds and high lubricating oil temperatures, at operating conditions under which the babbitt, tin-lead or related bearing metals fail. Without the use of lead segregation suppressors, the manufacturing practice of the copper-lead materials is very difficult and unsatisfactory.

Calcium In Iron and Steel

Metallic calcium in the form of briquettes with sponge iron has been used as a deoxidizing agent for iron castings made in green sand molds.[57] 0.5 per cent calcium is used, and the calcium-treated iron tends to be denser and more uniform in grain structure than ordinary cast iron and to have increased transverse and tensile strength and better impact values. Further work developed the use of calcium metal alone, additions of which to molten cast iron aided in controlling the amount and coarseness of graphitic carbon, and increased the strength of the casting.[58] Comparative values for ordinary cast iron and calcium-treated cast iron as well as heat-treated cast iron show definite advantages due to calcium additions.[59]

Good results have been obtained with steel,[60] the calcium functioning as an effective deoxidizer and degasifier, thus producing cleaner steels. It has the advantage over other deoxidizers that practically none of the reagent is left in the metal. Improved characteristics of the steel are shown in higher yield points and greater tensile strength.

Methods had to be devised for introducing the calcium metal into the steel. Briquetting with sponge iron or steel chips gave only partial success. Mechanisms[61] were finally developed by means of which slugs or pieces of calcium were shot into the ladle of molten metal through an air pressure gun. Sufficient force can thus be applied to the calcium projectile to shoot it deep into the metal where it reacts, disintegrates, does its deoxidizing and is converted into lime which rapidly rises to the slag.

Originally, projectiles of specific shape were employed, but later modifications of the gun use irregular chunks of calcium weighing about 2 oz. each. At the present time slugs can be shot at the rate of 120 to 150 per min. when the gun is hand operated, or 200 to 300 per min. if it is motor driven, and a 60-ton ladle of steel deoxidized in a period of approximately 5 min. when 2 lb. of calcium per ton of steel is used.

One application has been the use of calcium in deoxidizing and cleaning cast iron used in chilled rolls on which steel mill products are formed. One pound of calcium to 5,000 lbs. of iron gave the same scleroscope hardness as 15 lbs. of molybdenum to the same weight of iron.

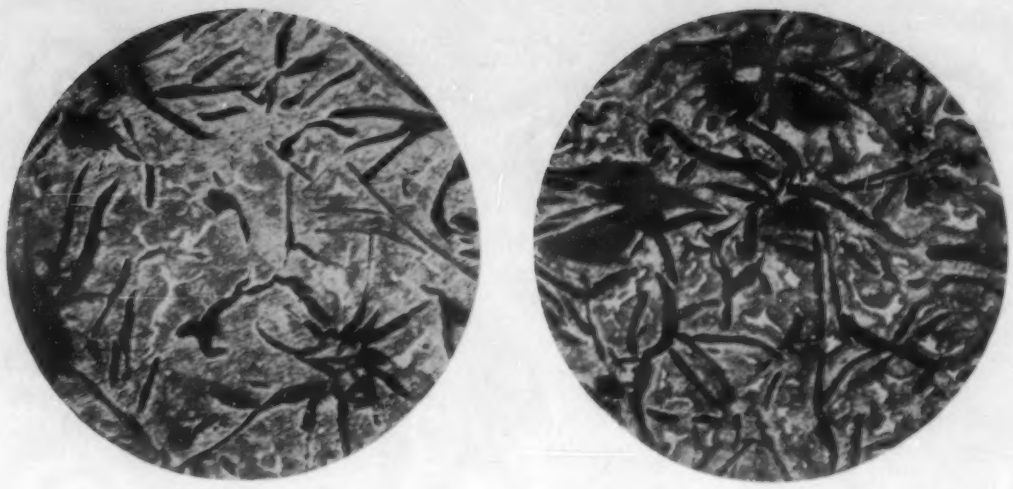
Calcium finds commercial application in the manufacture of the 50 nickel-50 iron alloys as a deoxidizer and scavenger. In pure nickel casting and rolling it appears to have some value in reducing internal defects such as seams, and improving malleability. The amount used is of the order of 0.005 to 0.5 per cent.[62] In the case of the chromium-nickel-iron alloys, it is regularly employed to prevent the formation of chromium carbide. Its use tends to prevent cracking, makes rolling less difficult, and aids when thin sections of the high-chromium high-nickel alloys are cast.

Considerable interest has been shown in the possibility of application of calcium to the making of chromium-nickel steels, particularly of the low carbon austenitic type, because it gives clean grain boundaries and uniform grain size, and retards the formation of carbides. The resulting cast forms show no mottling or clouding. These steels include both the ordinary and free-cutting 18 chromium 8 nickel with 0.1 per cent or less carbon; the 18 to 25 per cent chromium with 8 to 10 per cent nickel; the 25 to 31 per cent nickel steels with about 0.3 per cent carbon, 0.9 per cent Mn, and 0.35 per cent maximum silicon; the 34 to 36 per cent nickel steels with manganese of the order of 0.9 to 1.0; and the 41 to 43 per cent nickel steels. Other ferrous alloys in the manufacture of which calcium is used include the nickel-aluminum-iron alloys, (e.g. 29 nickel, 11 aluminum, 0.9 to 1.1 manganese) as well as a number of magnet steels. In the case of the high-chromium high-nickel iron alloys such as the 25 per cent chromium 15 per cent nickel with manganese of the order of 1.0 to 1.2, some operators claim that the use of calcium enables forgeable materials to be produced, and that if the calcium treatment be omitted, difficulties are met in forging. The use of calcium in nickel-chromium-iron resistance alloys is discussed below.

Calcium In Nickel-Chromium Alloys

In the nickel-chromium alloys and nickel-chromium-iron alloys used for high temperature electrical resistances, the oxides or sulphides, according to Hunter,[63] that collect along the grain boundaries are points of attack, and their elimination by the use of a deoxidizer such as calcium, with the retention of definite amounts of the deoxidizer, markedly improves the useful life of these alloys. Hunter specifically refers to alloys of the 80 per cent nickel, 20 per cent chromium type, with 0.03 to 0.2 per cent calcium.

Lohr[64] in a number of patents claims that the use of calcium with other elements, particularly zirconium and aluminum, markedly increases their useful life, as determined by the standard American Society of Testing Materials procedure for high resistance, high temperature alloys. The analyses of the alloys covered by these patents are given in Table III.



Polished and Etched Specimens of Ordinary (left) and 0.1% Calcium-Treated (right) Cast Iron. The black areas are graphite, the half tone areas are pearlite and the white areas are ferrite and phosphide eutectic. The calcium-treated sample shows less decarburization of the pearlite immediately surrounding the graphite flakes.

Calcium In Aluminum, Magnesium and Beryllium

For a considerable period, calcium has been used in connection with the preparation of aluminum and its alloys. In the early development of the strong alloys, calcium was a useful addition in materials which were to be forged or mechanically shaped. In the aluminum alloys containing copper and magnesium with or without manganese, calcium eliminated hot shortness during rolling of the ingot into sheet.[65] The amount of calcium added is 2 per cent or less, generally of the order of 0.6 to 1 per cent. In Europe and England, calcium is used in aluminum containing silicon to precipitate the silicon out of solution and form calcium silicide, with resultant elimination of the deleterious effects of silicon on the electrical conductivity of the aluminum.

Aluminum alloys containing about 4 per cent copper and 0.5 per cent magnesium[65] give a great deal of trouble and much scrap during the hot rolling of the ingot into slabs, because of the tendency of the metal to crack very badly. The cracks extend from the edge of the slab into the body and necessitate extensive shearing. These alloys are greatly improved by the addition of calcium in amounts of 2 per cent or less. In a number of aluminum-manganese-magnesium alloys, 0.3 to 0.6 per cent of calcium is suggested.

The modification of silicon-aluminum alloys by the addition of calcium and substances of similar properties has been investigated.[66] Examination of the microstructures of the normal and modified alloys shows that calcium and the alkali metals, sodium, potassium and lithium are very effective as modifying agents. The theory that modification is brought about by volatile elements is disproved by the fact that phosphorus, sulphur, selenium, tellurium, and arsenic give no appreciable effect.

Beryllium oxides (with iron) can be reduced to form beryllium-iron alloys only with calcium, or reducing agents containing calcium.[67] Magnesium, aluminum, silicon, lithium and cerium did not so function. Iron-beryllium alloys with up to 10 per cent beryllium may be formed by reacting beryllium oxide with a magnesium-calcium alloy containing 40 to 60 per cent calcium in the presence of Fe_2O_3. The beryllium and iron compounds are reduced and alloyed.[68] Boiling magnesium does not dissolve beryllium, but in boiling calcium an alloy of approximately 71 per cent beryllium and 27 per cent calcium is formed.[69]

The addition of 0.01-2.0 per cent calcium and 0.05-0.4 per cent antimony or bismuth, or both, to aluminum alloys that contain 2.0-15.0 per cent magnesium is said to be beneficial.[70] Calcium may likewise be added to improve such alloys when they contain, in addition to magnesium and antimony or bismuth, or all three, one or more of such alloying elements as cobalt, copper, nickel, manganese, zinc, etc., which may be added to modify or produce a specific property of the alloy. The addition of 0.01 to 2.0 per cent calcium and 0.05 to 0.4 per cent antimony or bismuth, or both, is very beneficial in the case of a series of aluminum alloys which contain

Table III—Composition, Per Cent

Patent	Nickel	Chromium	Iron	Molybdenum	Zirconium	Aluminum	Calcium
2,005,430	Balance	15–25	....	1–20	0.1–1.0	....	0.01–0.20
2,005,430	Balance	15–25	....	2–10	0.1–0.3	....	0.02–0.05
2,005,430	Balance	20	....	5	0.2	....	0.02
2,005,431	Balance	15–25	....	....	0.01–0.50	0.01–1	0.01–0.20
2,005,431	Balance	15–25	....	....	0.20	0.07–0.38	0.03
2,005,432	Balance	10–15	25–30	1–20	0.1–1.0	....	0.01–0.20
2,005,432	Balance	10–15	25–30	2–10	0.1–0.3	....	0.02–0.05
2,005,432	Balance	15	25	5	0.2	....	0.02
2,005,432	Balance	10–18	17–30	2–10	0.1–0.3	....	0.02–0.05
2,005,432	Balance	10–18	17–30	1–20	0.1–1.0	....	0.01–0.2
2,005,433	Balance	10–15	25–30	....	0.01–0.50	0.01–1.0	0.01–0.20
2,005,433	Balance	10–15	25–30	....	0.20	0.07–0.38	0.03
2,005,433	Balance	10–18	17–30	....	0.01–0.50	0.01–0.10	0.01–0.20

as major alloying elements 2.0 to 10.0 per cent magnesium and 0.2 to 5.0 per cent nickel. A useful aluminum alloy contains 3.0 to 8.0 magnesium, 0.5 to 4.0 nickel, 0.5 to 4.0 manganese, 0.01 to 2.0 calcium and 0.05 to 0.4 per cent antimony or bismuth, or both.

The ternary alloys of magnesium, zinc and calcium have been investigated.[71] The compounds in the system include Ca_3Mg_5, Ca_5Zn_2, and $Ca_2Mg_5Zn_5$. The last exists as large polygonal crystals, insoluble in nitric acid, melting without decomposition at 495 deg. C. Alloys containing 30 per cent or more of calcium oxidize readily; alloys rich in zinc oxidize slightly, are hard, break easily and are difficult to work; alloys rich in magnesium are light, oxidize but slightly and are easily worked.

Calcium In Precious Metals

An X-ray study of the silver-calcium system[72] gave no indication of the compounds $AgCa_2$, Ag_2Ca or Ag_4Ca. The compounds AgCa (face-centered cubic with a = 9.07 A. U.) and Ag_3Ca (tetragonal with c/a = 0.88 and c = 9.96 A. U.) were found.

A thermal investigation of the gold-calcium system[73] showed the existence of six intermediate crystal forms: Au_4Ca, Au_2Ca, $AuCa_{1.11}$, Au_3Ca, $AuCa_{1.33}$ and $AuCa_2$. The phase $AuCa_{1.11}$ has an extended range of homogeneity, from 49 to 56 atomic per cent calcium.

The deoxidation of commercial silver alloys was studied[74] by determining the amount of oxygen present after deoxidation and by microscopic and macroscopic examination of the deoxidized alloys. The use of calcium, magnesium, lithium, beryllium, aluminum, silicon, manganese, boron, zinc, phosphorus, tin and cadmium was investigated in preliminary experiments. The elements are considered in two groups: The oxides of the elements from calcium to boron in the above list have a heat of formation greater than 45 kg.-cal. per equivalent and the elements can react completely with an equivalent amount of oxygen; the last four elements are only partial deoxidizers. In the first group the deoxidation rate is determined by the melting point of the elements. Those with a melting point below 1000 deg C. react nearly completely in 1 min. If the melting point is from 1000 to 1500 deg. C., deoxidation requires from 3 to 5 min. If the melting point is very high, solution of the deoxidizer is slow and the reaction is incomplete even after 5 min. Calcium was found to have commercial possibilities.

Conclusion

It is difficult to tell how much of the above that is taken from patent literature represents actual and useful application—much of it has to be cited merely on the say-so of the patentee. A reasonable proportion will probably be substantiated and used as experimentation goes on. New metallurgical applications of calcium are constantly in the development stage or reaching commercial usage. Increasing knowledge emphasizes the specific properties, and their special utility and effects, as contrasted with those of the alkaline earths, the alkali metals, or the materials thought of in the general class of scavengers, deoxidizers and purifiers.

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letters

TO THE EDITOR

Nomenclature Again

To the Editor: I have been following with much interest the pertinent controversy between Dr. Sauveur and Messrs. Bain, Davenport, Villela et al. on the subject of the nomenclature for the constituents of steel. I should like to call attention to one aspect of the controversy which has not been mentioned, and that is the use of the nomenclature in the classroom and school laboratory.

The average beginning student in iron and steel seems to be in a state of mild confusion over the new terms he must learn by name, and over phases he must identify under the microscope. When he is introduced to the subject of heat treatment, his confusion increases enormously when he is confronted by the apparent vagueness of some of the terms as they are used in the older nomenclature, such as lamellar sorbite, lamellar troostite, granular sorbite and granular troostite, and his inability to correctly apply them to the structures he observes. This confusion in the student's mind is not completely cleared up until some time after he finishes the course and acquires more experience either in industry or through graduate work.

On the other hand, the nomenclature proposed by Villela and Cooper (METALS AND ALLOYS, Sept. 1938, p. 223) is simpler, more logical and is readily accepted by the student because he can easily apply it to the structures he observes under the microscope without the necessity of considerable coaching by the instructor.

The 32 metallurgists to whom Dr. Sauveur submitted his questionnaire are without doubt authorities on the subject, but this very fact means that they are men who have used the older nomenclature for many years and hence for that reason may be very reluctant to give it up in favor of a newer, simpler terminology. In support of my contention, I should like to see Dr. Sauveur's questionnaire submitted to 32 younger metallurgists between the ages of 25 and 30.

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Inclusions in Cast Iron

To the Editor: In the article on "The Nature of Inclusions in Cast Iron" by Profs. Scott and Joseph in the November issue of METALS AND ALLOYS, some excellent photomicrographs are shown in which gray inclusions are clearly evident. The discussion of these photomicrographs indicates that the authors were in doubt as to the nature of the inclusions illustrated, and contemplated using petrographic methods for identification. It would be interesting, at least to the writer, to learn why the well-known metallographic tests were not applied to these samples to determine the nature of the inclusions shown. By means of theses tests sulphides, silicates, alumina and titanium carbide or cyanonitride are nearly always readily distinguishable, and it would seem that this should be an obvious routine procedure in any work on the nature of inclusions in steel or iron.

Although it is impossible to judge accurately from an unetched photomicrograph, the writer would be inclined to hazard a guess that nearly all the gray inclusions shown in Profs. Scott and Joseph's photomicrographs are sulphides. These are generally the most common inclusions seen in polished sections of cast iron, as they are easy to show clearly with ordinary methods of polishing. Of course, this does not mean that sand grains may not also be found.

The writer would recommend to the authors that more attention be paid in this study to the prevalence of sulphide inclusions in cast iron; and that titanium cyanonitride, which can be seen in well-polished microsections of most cast irons, be likewise considered in regard to its action in the form of nuclei during solidification, instead of concentrating the discussion almost exclusively on oxide inclusions.

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Discussion by the Authors

In view of the small amounts of oxides found by electrolytic extraction, the larger and most numerous ones are likely sulphides. The inclusions were so small that it did not appear feasible to attempt identification with etch tests. While such tests can be used to identify or classify inclusions as to type, they do not give sufficient information as to composition. Additional methods of study appear necessary to determine the origin and effect of inclusions.

Definition of Steels

To the EDITOR: The article in your December number on "Rimmed vs. Killed Steel" is full of interest, but, I think, starts wrong with its definitions of steels. So I send you mine. Then I find no mention of the surface defects common in killed steel.

Definitions of killed, semi-killed, and rimming steels are properly based on the control of the gases they contain. Killing and deoxidizing steel are not synonymous terms. To prevent the escape of hydrogen is not a matter of deoxidation. Steel may be killed without being deoxidized, as I have often stated. Killed steel may be redshort due to contained oxides. The gases of steel offer a great field for research, beginning not later than with the gases in crude iron.

Killed Steel is that which contains enough gas solvent to prevent, during freezing, escape, from the metal, of gas, which might otherwise form gasholes within it.

Semi-killed Steel is a misnomer. Rarely, if ever, is steel half killed. It is often partly killed, but the degree of killing may be any from a little to nearly total. No steel could be authoritatively declared to be half-killed.

Rimming Steel is low carbon simple steel which has boiled in the furnace, and to which, at the end, but little or no gas solvent is added. In the mold it evolves gas freely which keeps the liquid metal in lively motion until it freezes, which of course it does from the outside in. The solid shell, as it forms, shows around the top as a rim. Hence the name. HENRY D. HIBBARD.

Radiographing Heavy Cast Steel

To the Editor: I believe the following information will interest some of your readers.

As far as we know, or any one whom we consulted before we undertook this job, this is the first time that cast steel 9½ in. thick was radiographed. We used 500 milligrams at 15-in. focus to film distance for a period of 5 days, which is in accordance with the charts given in our book on "Industrial Radiography," but less than one-third of the time recommended by the Navy men. The actual area covered satisfactorily was about 10 in. in diameter or ⅔ the focal-film distance. This is due to the increased thickness to be penetrated by the rays toward the ends. A penetrameter 3/16 in. thick was placed on the casting side facing the radium and another one of the same thickness was attached to the film holder.

Only the penetrameter nearest the film showed up in the gammagraph. The regular photographic procedure recommended for this type of work was followed: 2 DuPont X-ray films between 3 foils of 0.06 in. lead, developed with potassium iodide at slightly over 70 deg. F. for 10 mins.; this is a 10 per cent increase over the regular developing temperature.

Besides the above experience which we gained for the first time, we learned that a larger amount of radium is much more effective than has been taken for granted by workers who arrived at their statements after using smaller amounts of radium only. This was particularly noticeable when we checked the protective efficiency of the container. While this very same container may be perfectly safe for transporting 100 mg. of radium, it is absolutely unsafe for the use of 500 mg. We have such evidence in our files and again emphasize the necessity of proper protective devices as recommended by the writer in METALS AND ALLOYS, April, 1935, page 100. The hidden dangers involved in frequent handling of radium materials cannot be emphasized often enough and this material should only be used with extreme care.

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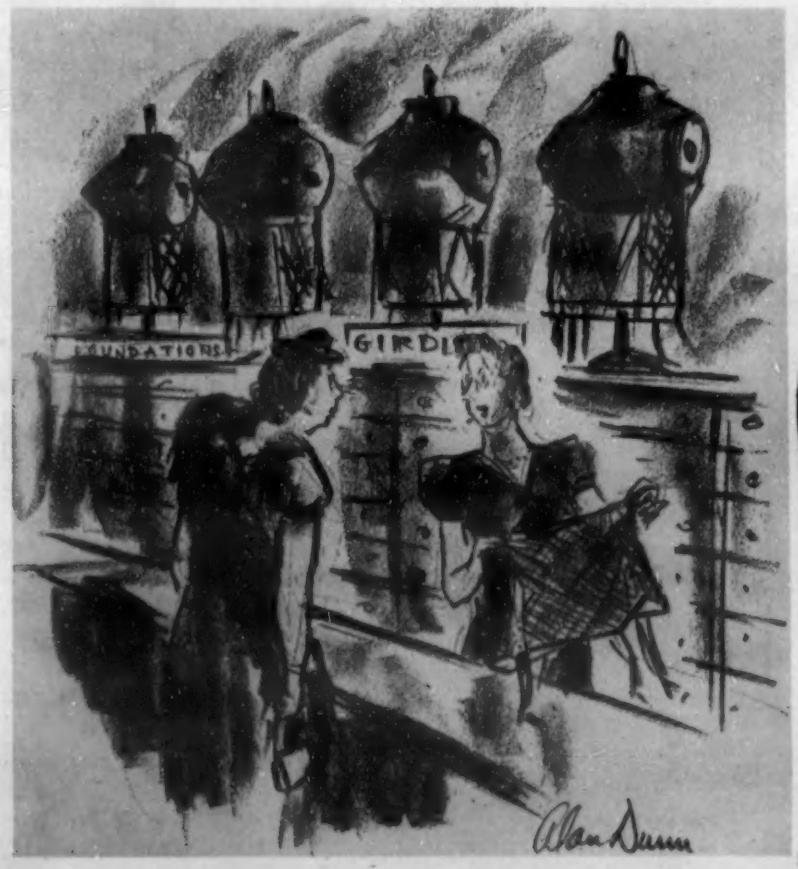
A Couple of Chuckles

Non-Metallurgical "Creep"

Our secretary to the editor, because of her association for several years in a metallurgical atmosphere, has an eye for the "chuckle-i-ferrous" in metallurgy. She believes that the illustration here reproduced will interest the members

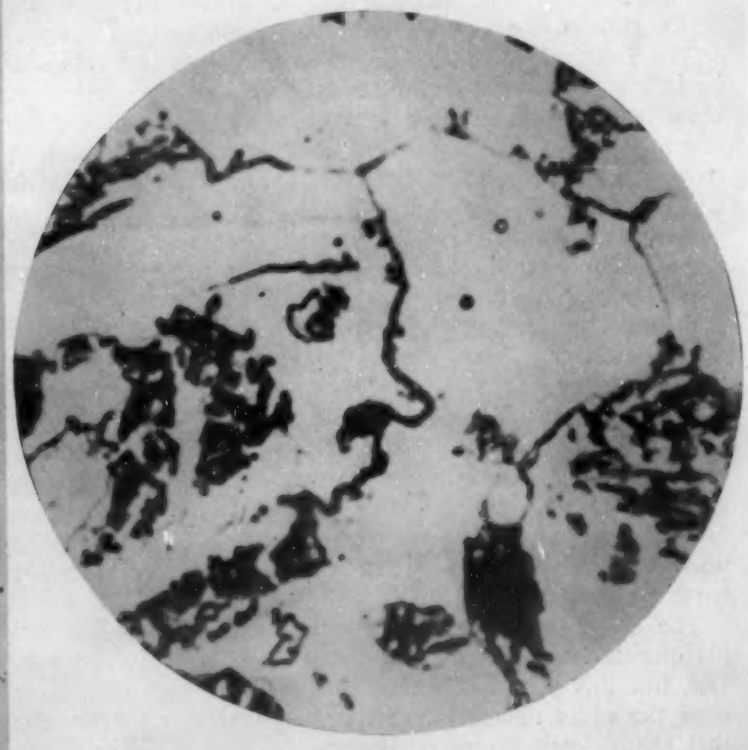


of the high temperature committee as well as metallurgists in general. She spied this in *The New Yorker,* for Sept. 3, 1938, and recommended it to the editor as a chuckle. He seconds the motion.—E.F.C.

The "Body" of Steel

In your issue for December, 1937, page 361, you published a micrograph to substantiate the belief in the "body" of steels.

If further evidence of the "personality" of some steels is



still necessary, surely the untouched micrograph sent herewith at a magnification of 2000 diameters will convince the most incredulous of your many readers.

J. H. SLAGG.

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